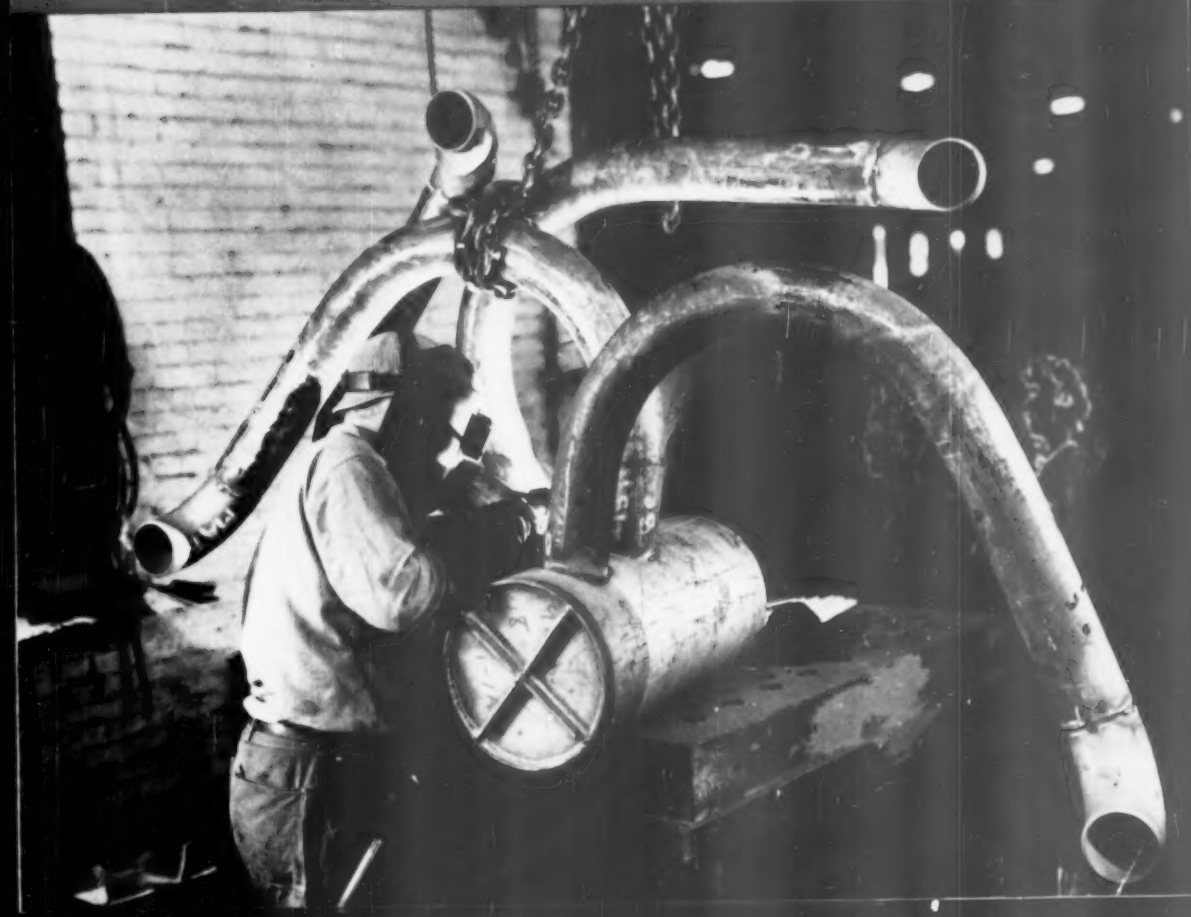
JULY, 1950

The WELDING JOURNAL

Official Publication of the American Welding Society



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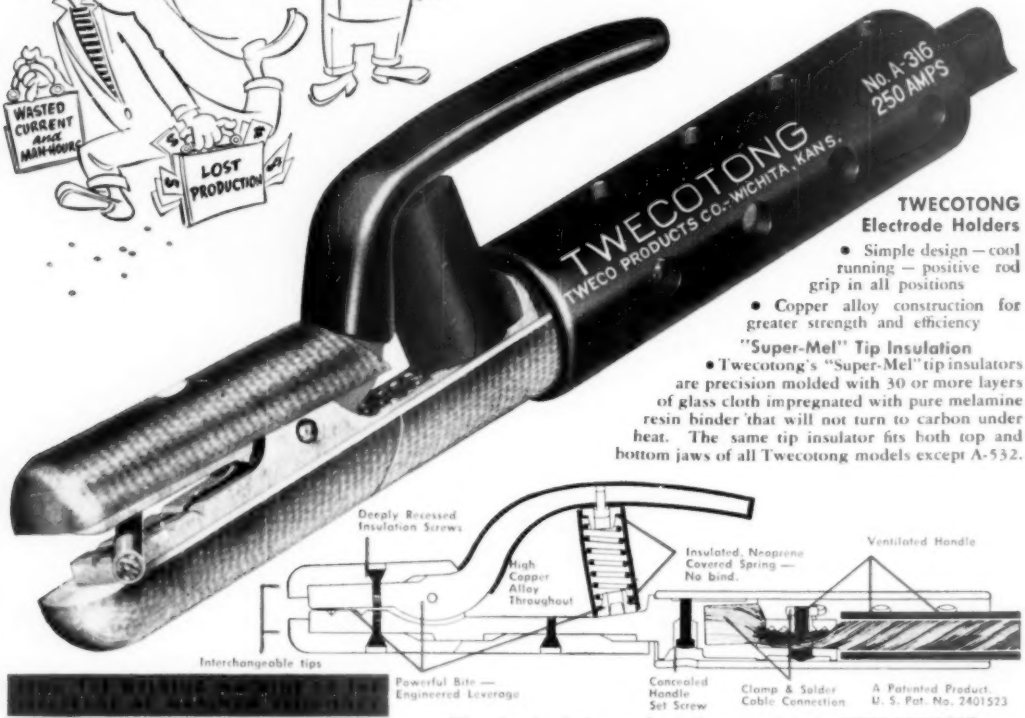

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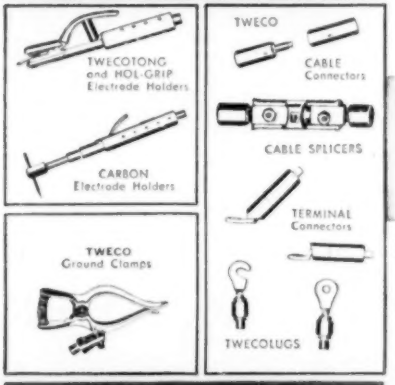

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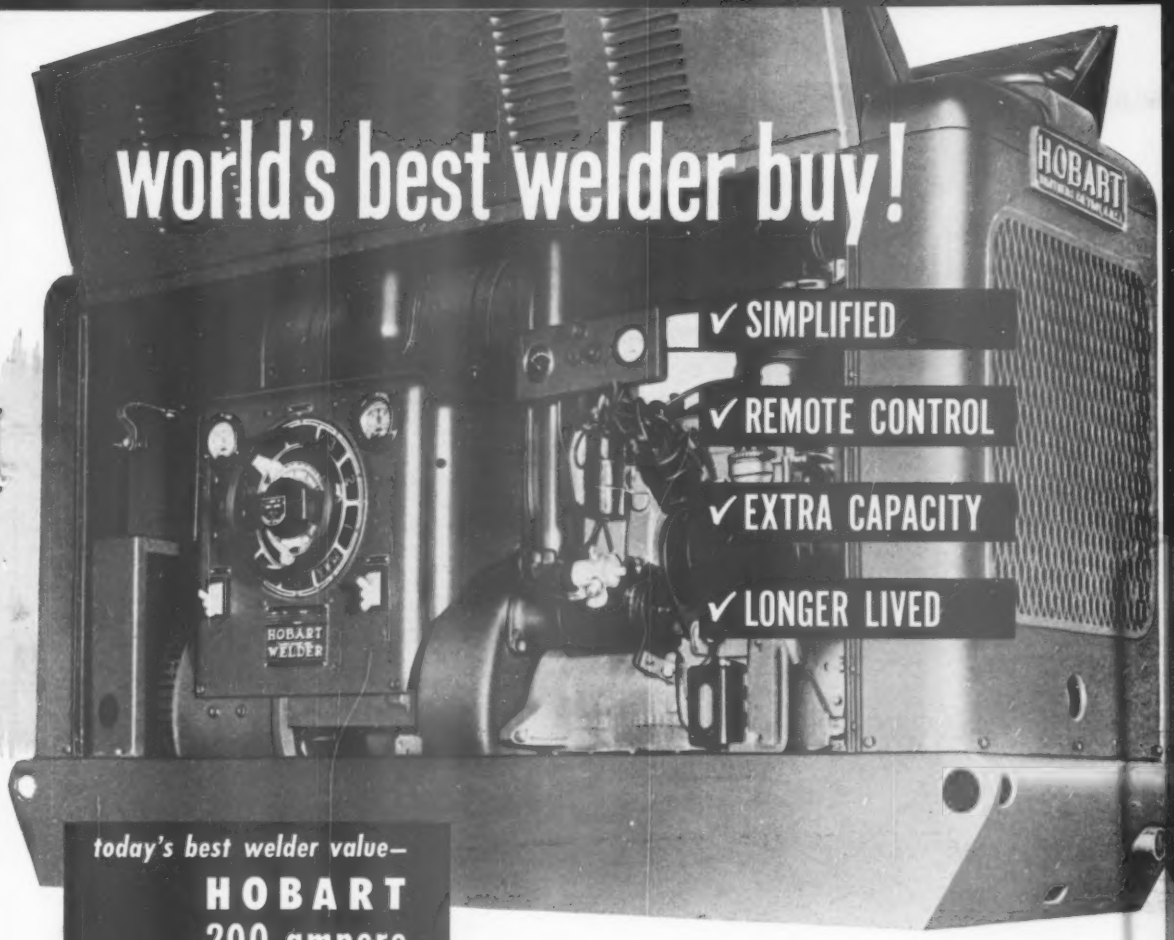

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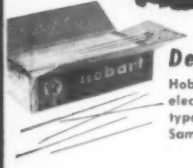

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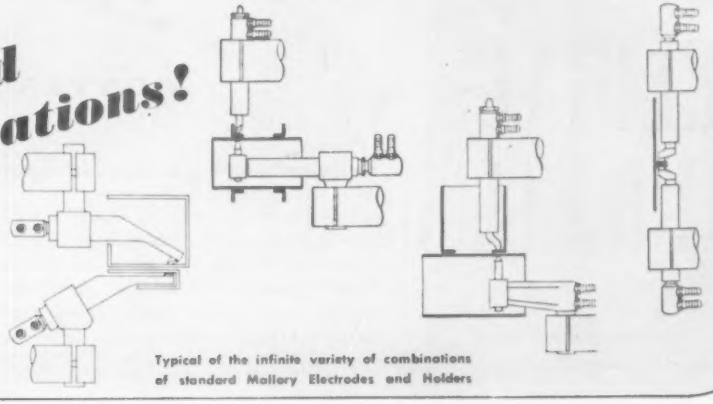

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cover: An example of a complex assembly in stainless steel with accuracy maintained in every step of production. Courtesy Benjamin F. Shaw Company.

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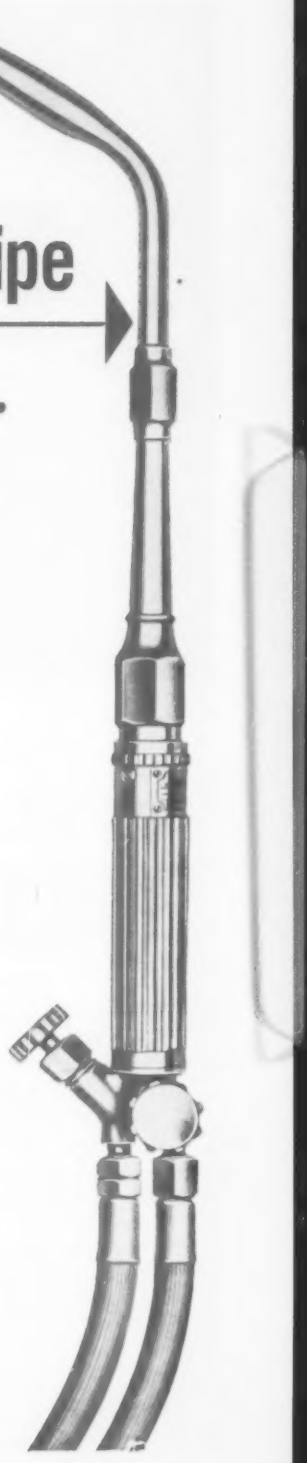

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The WELDING JOURNAL

volume 29 July 1950 number 7

Applications for Helium in Inert-Arc Welding

An interesting historical review stressing at the same time the advantages of helium in inert-gas-shielded arc welding and its promising future

by M. J. Conway

SUMMARY

HELIUM is one of the two monatomic gases which are practicable for inert-arc welding. The other is argon. Neon, krypton, xenon and radon are too rare and expensive.

Helium is not widely used in many applicable cases. Its use is preferable in welding stainless steel, inconel, Monel, cupraloy, silicon bronze and copper in all but the thinner gages.

At present, direct-current straight polarity is used most with helium; because it gives a more inherently stable and hotter arc.

Alternating current is used but little with helium at present; because higher voltage is required for arc stability, there is less arc heat per ampere, and conditions are more critical.

Henry M. Hobart of General Electric corresponded with associates about helium in welding in 1919, and he received one of the basic patents on inert-gas welding in 1930. However, commercial development of welding with helium was delayed until World War II, when it was used in welding magnesium and stainless steel.

Postwar development has been rapid. Helium was found preferable for welding some other difficult metals. The U. S. Bureau of Mines cooperated by constantly increasing helium purity to its present 99.99%, and by undertaking some of the practical welding research. Refinements of alternating current circuits give great promise, especially in welding aluminum. There is at least one application where helium has proved decidedly superior to argon in welding aluminum. Welding of other metals with helium as the inert gas and with either alternating or direct current sources is being investigated.

There is little doubt that a much wider application of helium in inert-arc welding is in the foreseeable future.

INTRODUCTION

The fact that helium is monatomic is important because it shows how exclusive, i.e., inert it is. It will not even unite with a fellow atom to make a molecule, as all active gases such as oxygen and hydrogen do, let alone combine with other kinds of atoms to form chemical compounds.

This is true even on the supertorrid surface of our sun. There, even hydrogen atoms undergo fission, thereby releasing the energy we get from the sun. The split hydrogen atom parts reunite into helium atoms. And there the action stops. The resultant helium will not break down or unite with anything on the sun, in spite of the tremendous heat and energy there.

It follows then that helium certainly will not break down or unite with anything under the influence of a welding arc. This makes it an ideal inert welding atmosphere.

While it is true that this inertness is shared by the five other monatomic gases—argon, neon, krypton, xenon, and radon—only argon approaches helium in relatively cheap and plentiful supply. When and if the others become even nearly so, their characteristics as welding atmospheres will no doubt be investigated.

The use of argon is too well known to require further comment here, except to dispel one pseudoscientific bit of fiction which has sometimes mitigated the use of helium. This fiction is that argon, being heavier than helium, even heavier than air, will lay down around and blanket the arc and molten metal better than helium. The facts are these: Argon is not significantly heavier than air; if it were it would displace the air and blanket the earth's surface to well over the average

M. J. Conway is Application Engineer, General Electric Co., Fitchburg, Mass.

Presented at the Thirtieth Annual Meeting, A.W.S., Cleveland, Ohio, week of Oct. 17, 1949.

height of man. Argon does have an atomic weight nearly ten times that of helium (argon, 39.88; helium, 4.00). However, when heated to over 3000° C. by a welding arc, both gases tend to rise at a rapid rate. Shadowgraphs show little, if any, difference in this rate for argon or helium. Only the velocity of the gases and the flow pattern produced by the contour of the gas nozzles, or cups, seem to be effective factors in gas coverage. Argon and air are about the same weight, and we all know how even slightly hotter air rises. To re-express the point, hot gases rise, and the hotter the faster. Until we get working with radon, whose atomic weight exceeds that of mercury, the heaviness of the inert gas used in welding would seem insignificant.

PRESENT APPLICATIONS

Stainless Steel

The widest use of helium today is in the inert-arc welding of stainless steels. There are several good reasons: When used with a direct-current, straight-polarity source of welding power, welding speeds increase as much as 40% when helium is substituted for argon. Again, this speed is as much as 50% greater than when argon is used with an alternating current welding power source.

There is nothing mysterious about the attainment of these higher welding speeds. With helium the arc voltage per given arc length is at least 40% greater than with argon. Therefore, the wattage, or heat, in the arc is that much greater. Since the speed of welding is a function of the rate of heat input to the work, it follows that the hotter arc attained with helium produces faster welding. Only where the work is too thin or too small to take the higher heat of helium is the use of argon and direct current preferable for inert-arc welding of stainless steels.

Other advantages accrue with the higher welding speed when helium is used. Distortion is reduced both directly because of the increased welding speed and also because of the reduction in width of the heat-affected zone and the weld bead itself. The narrower weld bead and heat-affected zone plus the faster fusion and chilling rates, decrease carbide precipitation and other undesirable metallurgical changes.

The use of reverse-polarity direct current to reduce heat input for small or thin work has, unfortunately, proved impractical. The reasons are: first, an extremely short arc length is required to maintain an arc; and second, the arc tends to wander over a fairly wide surface in a glow rather than a solid welding arc. This observation appears to be generally true whether helium or argon is used, and with other metals. The notable exception is magnesium, which welds nicely with helium and reverse-polarity direct current. This wide brush-type, unstable arc appears to be due to the electrical emission characteristics of various metals; i.e., their ability to give off electrons to form an arc when they are negative and a positive electric pole or point is brought near. It seems that most metals give off electrons, or emit, more readily than tungsten. Therefore, when the tungsten electrode is made positive, as it is in reverse polarity, it tends to pull the easily moved electrons off a wider surface of the electrically negative work. Moreover, certain small points on the workpiece are sufficiently good emitters that any solid part of the arc jumps from them rather than from the point directly under the electrode. These high emission points are called cathode spots. It is the arc's tendency to jump around to these cathode spots that makes it unstable. Of course, an extremely short distance between the electrode and the work limits this effect by narrowing the effected area.

The work done to date in using helium with an alternating current source to weld stainless steel has given inconclusive results.

Inconel and Monel

These two metals are also often welded with helium and direct-current straight polarity. Welding conditions and comments are similar to those for stainless steel. The only additional point worthy of attention is the fact that X-ray quality welds can be made in inconel and Monel with the inert-arc process. This remains true whether or not filler metal is added.

Copper and High-Copper Alloys

Here again helium has proved preferable to argon principally due to the hotter arc obtained with helium. This hotter arc requires less preheat, lower amperage welding current (often reducing the required equipment capacity) and gives the expected higher welding speed and reduced distortion.

Direct-current, straight-polarity power is preferred because in addition to a stable arc it gives the best division of arc heat—the majority going to the work, and comparatively less to the electrode. When copper's high capacity to absorb heat and conduct it rapidly away from the weld point are considered, the advantages of this greater arc heat input to the work becomes apparent.

Although some moderate success in welding copper and high-copper alloys has been obtained with argon and alternating current, results with helium and alternating current are inconclusive.

Welds in silicon bronze made by the inert-arc process, with helium as the inert gas, will meet A.S.M.E. Code U-68 for Unfired Pressure Vessels.

When inert-arc welding copper or high-copper alloys, it is preferable to weld uphill some 10 to 15° whenever possible. Otherwise, there is a tendency for the very fluid metal to flow ahead of the arc. This results in poor fusion and difficulty in laying in relatively heavy passes.

Brasses are weldable with helium if some zinc and color loss are permissible.

Aluminum

Helium is not used to any great extent in inert-arc welding of aluminum.

Helium can be used with direct-current straight polarity if flux is used. However, as soon as flux is used the chief advantages of inert-arc welding—freedom from flux handling and removal, possible flux inclusions and corrosion, and porosity—are lost. Joint design must then provide access for flux removal and neutralization to prevent corrosion. Even with straight polarity and flux, however, argon is still preferable for inert-arc welding aluminum.

Reverse-polarity direct-current welding of aluminum with this process has not proved practical under production conditions.

When using alternating current in inert-arc aluminum welding, helium has proved preferable to argon in making plug welds. (Plug welds are invaluable in permanently fastening aluminum strips or sheets to heavier aluminum structural members or bus bars.) The reason helium is preferable in this case is that the arc shows less tendency to burn back the edge of the hole in the thinner strip than when argon is used. This means that a neat fillet can be run around the edge of the hole before filler metal is added to complete the weld. This fact strongly suggests that helium and a. c. might be preferable to argon and a. c. in making lap welds in aluminum. This theory, however, is untested to date.

Helium and alternating current has consistently produced outside corner and edge joints in aluminum comparable to those produced with the more usual alternating current and argon. Work on other joint types in aluminum is not conclusive. For instance, practically no aluminum work involving filler metal addition has been done with helium to date.

Hard Facing

Some hard facing is now being done by the inert-arc process employing argon and straight-polarity direct current. This has proved equally satisfactory to and several times faster than gas welding for this purpose. The hotter helium-shielded arc may reduce the welding time required still more—that remains to be seen.

It must be apparent from the foregoing that although the use of helium with direct current is fairly well established in inert-arc welding, the helium and a.-c. combination is in the same developmental stage as was argon and alternating current five years ago. The development of the argon a.-c. combination in inert-arc welding is mostly history.

HELIUM HISTORY

Speaking of history, let us take a peek at the letter Henry M. Hobart of General Electric is reading—the date is March 7, 1919:

"Dear Mr. Hobart:

I have received your two recent letters regarding the use of helium gas in connection with arc welding.

I believe your scheme would work well and improve the quality of welding. At any rate, it is well worth trying. . . ."

However, basic research takes time and helium was hard to get and costly. So it was not until 1930 that Mr. Hobart was to receive and assign to the G-E Company the basic patent on inert-gas welding.

World War II arrived before development made the inert-arc process practicable. By then helium was being used with reverse-polarity direct current to weld magnesium aircraft parts. As the war progressed, helium and straight-polarity direct current were used for stainless steel and for aluminum welding with flux. However, before the War's end, aluminum was being welded without flux using argon and alternating current.

After the war, intensive development of inert-arc welding was stimulated by three factors.

1. The steel shortage produced demands for better nonferrous metal fabrication methods.
2. Engineers and scientists were able to follow through on further theories and equipment development.
3. The U.S. Bureau of Mines, with a surplus of helium on hand, became interested.

The interest of the Bureau of Mines led first to cooperative investigation with Industry and then to the acquisition of equipment and further investigation in their own laboratories. This led to increasing the purity of helium further and further until today a purity of 99.99% has been attained.

This increased purity has proved worth while by decreasing and even eliminating many previous problems. (One of these was the elimination of porosity in inert-arc welded Monel.)

As a result of this developmental work by the Bureau of Mines, helium of purity better than 99.9% or double-charcoal helium, is being furnished as standard for inert-arc welding.

FUTURE

The work the U. S. Bureau of Mines is doing in developing an a.-c. helium-aluminum welding technique looks very promising. Anyone who has seen samples resulting from this work would find it difficult in some cases to tell them from those made with the now standardized argon a.-c. process. Since aluminum has been one of the most difficult metals to weld by a fluxless process, it follows that perfection of an a.-c. helium technique for aluminum would undoubtedly open new fields of application for other metals.

In a recent issue of THE WELDING JOURNAL an article appeared which indicated that Air Reduction com-

pany's work in developing gas-shielded metallic-arc welding was aided by the availability of double-charcoal helium. There is an excellent chance that helium may replace argon in this application for aluminum welding. In turn, the perfection of an aluminum-welding technique will allow the extension of development of the gas-shielded metallic-arc welding process to other metals. Surely, helium will be found preferable for some of these metals.

This is only one indication of the applied research being carried on both by the welding equipment industries and the metal fabricating industries. Some developmental work in gas-shielded welding is being carried on also by relatively disinterested organizations such as Battelle Memorial Institute.

All of this research and development certainly promises new uses for helium as its characteristics become well enough known to make further technique problems more predictable.

Equally certain is the continued interest of the U. S. Bureau of Mines as long as they are compelled to pump helium back into the ground in order to preserve it.

Helium at Amarillo, Tex., is as cheap as hydrogen and only 15% of the average delivered cost of argon. When helium is used in sufficient volume to reduce handling and shipping costs to nominal figures, a still greater use will be dictated by economy. Then, helium may regain the ascendency it had at the start of inert-arc welding.

In any event, some of the inert-gas welding problems that plague us today will undoubtedly be solved with the help of helium tomorrow.

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Portable Resistance-Welding Tools

Design and application of portable resistance-welding tools, where used advantageously, limitations, co-ordination of product and design

by Charles H. Davis, Jr.

PORTABLE RESISTANCE-WELDING TOOLS

HALF a million welds a day! That is the estimated production of portable resistance-welding tools at the Budd Co.'s Red Lion Plant where stainless steel railway cars and truck trailer components are manufactured. This is an astonishing figure, but when it is multiplied by the great number of plants that use portable resistance welding, the total becomes astronomical. Why is this process so widely used and what are its advantages? In an effort to answer these questions a discussion of the design and application of portable resistance-welding tools is in order.

Once the use of resistance welding has been decided upon, the question often arises as to whether the equipment should be of a stationary or portable nature. The use of portable equipment is indicated when:

1. The work and necessary fixture is too large to transport to a stationary machine.
2. Production is greater than normal job shop methods will allow.

Because portable resistance welding is used mainly by manufacturing plants using mass production techniques, the small manufacturer frequently assumes that his production is not sufficiently large to warrant the use of portable resistance-welding equipment. However, the advantages of the resistance-welding process using portable tools is available to the small as well as to the large manufacturer. One of the greatest advantages is the flexibility which can be obtained with a comparatively small amount of equipment.

DESIGN LIMITATIONS

A typical comment overheard in the shop one day was "this tool is too big, too heavy and bends too much."

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This short statement effectively presents all of the major problems encountered in the design of portable welding tools. The designer is necessarily limited by various factors, many of which are beyond his control.

Coordination of Product and Process Design

The first limitation and probably the greatest, is the geometry and size of the parts to be welded. In any manufacturing process the requirements of the principal method of fabrication must be satisfied to obtain the lowest manufacturing cost. In this case to accomplish low-cost production the design of the product and the necessary fixtures must be closely coordinated with the welding-tool requirements. To this end it is desirable to have product designs and the jig and fixture designs reviewed in their earliest stages, by the welding-tool designer or welding-process man. This practice eliminates expensive changes when the final drawings have been issued to the shop for production, and also allows the most efficient use of the portable welding tool.

This recalls an incident that occurred some years ago. The product was for naval use and as such required the utmost strength in addition to being light in weight. One of the sections was so small that it seemed impossible to design a welding tool to meet the space restrictions imposed. At a conference between the product designers and the welding-tool designers it was decided there was no alternative but to try to build a welding tool to fit the condition. The tool was designed and built using a high grade of heat-treated alloy steel with a very small copper insert to carry the welding current. This tool proved satisfactory and performed well even though it was originally thought to be impossible. In the meantime however, cracks were found in the weld nuggets due to peculiar characteristics of the materials being welded. For this reason the weld which had formerly been considered indispensable by the product engineers was deleted from the design. The foregoing incident forcefully demonstrates the necessity of keeping an open mind on problems in which coordination is necessary between the product-design engineer and the welding-tool designer.

Choice of Materials

The second limitation imposed on the designer is in the choice of materials from which to make the tool. This choice is influenced by the conditions under which the tool must be used. If the tool must be used without any support it should be made as light as possible, even though this type of construction may be more expensive than those ordinarily used when a tool is supported by balancers or other devices. This problem has led to the use of lightweight aluminum or magnesium alloys in the construction of welding tools. One problem encountered in this design is the difficulty of attaching a copper or copper-alloy electrode to the light alloys. This problem has not been completely solved up to the present time.

The third limitation imposed on the designer is based on the material of the parts to be welded. This immediately fixes the material which must be used for the welding tips, and in turn indicates the method to be used in attaching the welding tip to the other components of the tool. In the case of the RWMA Class A materials, a mechanical connection is necessary as the temperatures necessarily associated with the silver-soldering operation will soften the material. On the other hand, tips of RWMA Class B alloys should be attached to the tool by silver soldering, otherwise expensive material is wasted in accomplishing a mechanical connection.

OTHER CONSIDERATIONS AFFECTING DESIGN

From the foregoing analysis it is quite evident that the welding-tool designer must be a capable machine designer if the most efficient design of the welding tool is to be achieved. In addition to the limitations imposed by the nature of the parts to be welded and the limitations of the materials required in the design of the tool there are many other considerations which must be taken into account before the tool is finally designed and built. Some of these basic considerations are as follows:

(*a*) The need for water cooling of the tool and tips.
(*b*) The expected life of the tool.
(*c*) The type of tool required "C," "X" or a combination of the two types.
(*d*) The physical size of the tool and its effect on the selection of materials to be used in its construction.

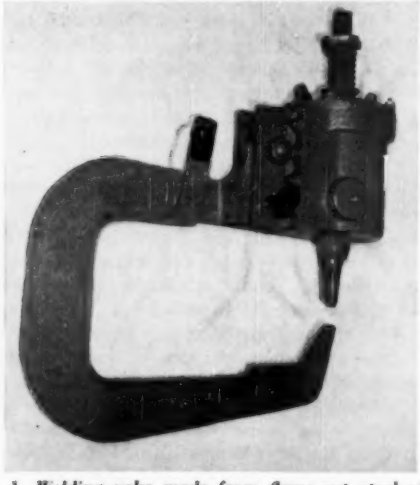


Fig. 1 Welding yoke made from flame-cut steel plate with copper bar added to carry welding current

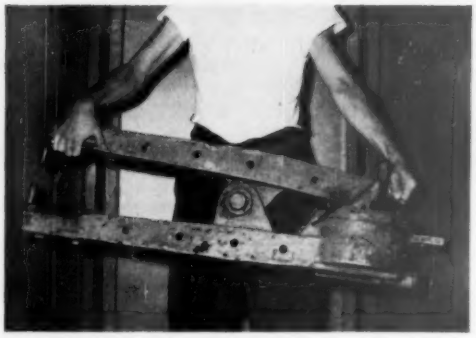


Fig. 2 Cast-aluminum X or scissors-type tool
Aluminum is used to carry welding current as well as physical load

Water Cooling

In the case of water cooling it is generally agreed that water cooling is necessary on high-production work or when the material to be welded is of heavy gage. It is much better to design a tool with water cooling than attempt to modify the tool by adding a water-cooling system after it is built. From the viewpoint of flexibility, a tool that is adequately water cooled is always capable of performing the highest quality welding within its physical limitations. Thus

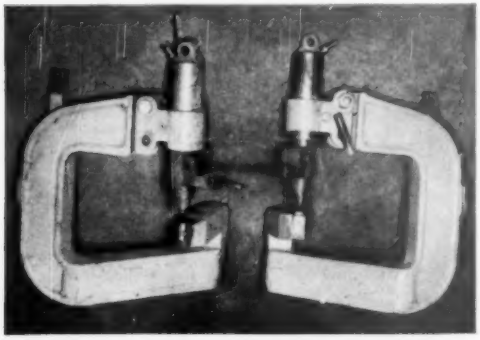


Fig. 3 Pair of hydraulic C-type tools fabricated from steel plate are welded to form box section. Tip force 3500 lb.

Fig. 4 C-type tool using a cutout lower electrode which may be rotated 360° without cable interference

the tool may be used in some future job where production requirements are more severe than those called for by its original work.

Life Expectancy

It may seem strange that the life expectancy would be considered in the design of a welding tool. The reason for this factor is quite evident when we realize that all materials have a fatigue limit. In the case of the welding tool a load is constantly applied and released during the entire life of the tool. This indicates that maximum design stresses must be based on the fatigue limit rather than the yield strength of the material.

It must be realized that if the space available for the welding tool is restricted we must expect to replace the tool after a certain number of operations. Since most of the fatigue data on cast material has been collected from polished specimens, tested as rotating beams, it is necessary to make further allowance for the possibility of surface irregularities or faults within the commercial casting. Therefore, the ultimate strength or the yield strength of any given material may be extremely misleading when the design stress of the tool is computed.

Type of Tool

Since the catalogs of most suppliers of portable welding equipment contain many examples of the different types of welding tools generally available, very little need be said about their selection. This selection is generally indicated by the nature of the work and the method of handling. There are cases, however, where none of the existing designs will fit the required application and it is then necessary to design a special single-purpose tool. Since a tool of this type has generally only one application, its cost including the additional cost of design will be very great and unless the anticipated production justifies the additional cost it is not economically feasible to use tools of this type.

Materials Used in Tool Construction

The physical size of the tool indicates immediately the choice of materials to be used in its construction. A welding tool must do two things, namely, apply force to the work and conduct welding current to the welding tips. At the present time we know of no single material that will perform both of these functions with maximum efficiency at the same time. Through the use of high-alloy steels we can achieve great strength, but these materials greatly restrict the flow of electric current. On the other hand, copper is the best-known conductor of electricity in general use. However, its physical properties are such that it is capable of carrying only a small portion of the mechanical load which an equal size section of alloy steel would carry. The best materials for welding tools available at this time are found in the RWMA Class A alloys. These materials combine a portion of the physical strength of steel with high-electrical conductivity and are available in either cast or wrought form. This material is expensive

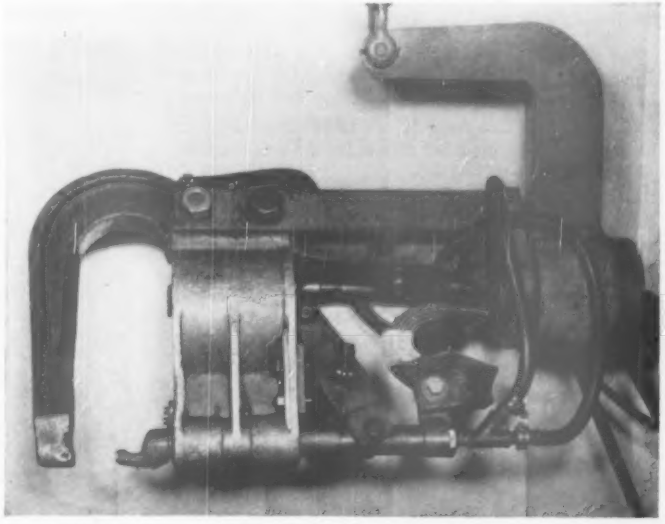


Fig. 5 Method of tool hanging used to eliminate cable interference as tool is rotated in job

Cable is led in through bearing hanger and twists with the tool

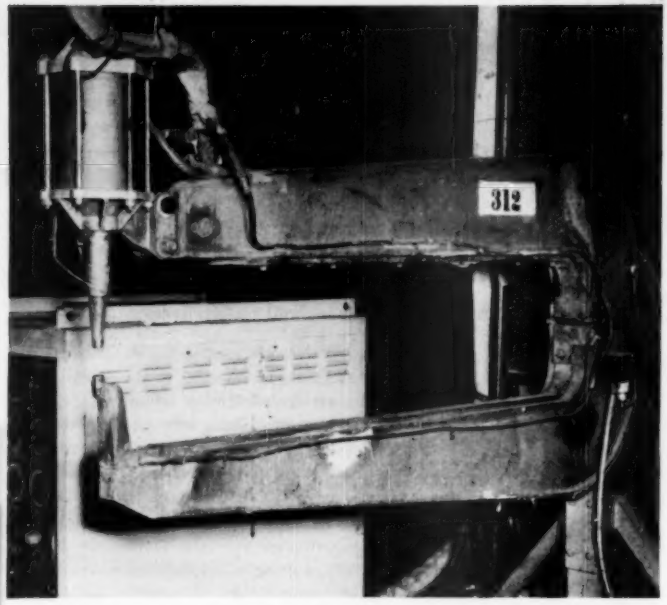


Fig. 6 Large C-type tool mounted on floor stand and used as a pedestal welder

when compared to steel of structural quality, and the limitations imposed by foundry practice make large castings expensive and difficult to procure. Generally speaking, a welding tool having a throat depth of more than 18 in. will be most efficient if designed around a structural steel member with a copper bar added to carry current to the welding electrodes.

Section Efficiency

The most efficient section should be used in the design of portable welding tools in order to limit the weight of the complete tool. An I-beam section will be found most efficient in this respect. In the case of castings this section can be modified to provide space for cast-in water tubes.

In larger yokes fabricated from steel plates, a box section is more efficient since it gives lateral as well as vertical stability. The danger point in each case is at the bend radii or at abrupt changes in section. Reinforcements are required at these points to prevent failure due to high local stresses. When tools of short throat depth are highly loaded, every effort should be made to avoid tool marks or other discontinuities in corners where stresses are normally greatest. The presence of any discontinuity at this point constitutes a stress raiser and is almost certain to cause premature failure

Deflection

It is to be noted that in tools of this size another factor must be considered in the design of the tool. This factor is the allowable deflection that can be tolerated without causing the welding electrodes to slide on the work. Generally speaking, the greater the throat depth of the tool the greater the allowable deflection. On some deep throat tools this deflection has been as much as $^5/_8$ in. On tools having extremely short throat depths it is possible to design to the fatigue limit for the material used, since the deflection obtained in such cases is negligible.

Cylinders

To those unfamiliar with the use of portable welding equipment it often seems strange that the welding tools are designed and built around air and hydraulic cylinders which are unlike those generally used for jig and fixture work. Most applications of cylinders in portable welding tools impose side thrust loads on the piston rod and bushings. The average stock cylinder is not designed for this loading and consequently will not stand up well under repeated use. Another factor affecting the cylinder design for welding tool use is that of weight, since it is desirable that the weight of the assembled tool be kept to a minimum.

TOOL SUPPORT

One item that should not be treated lightly is the

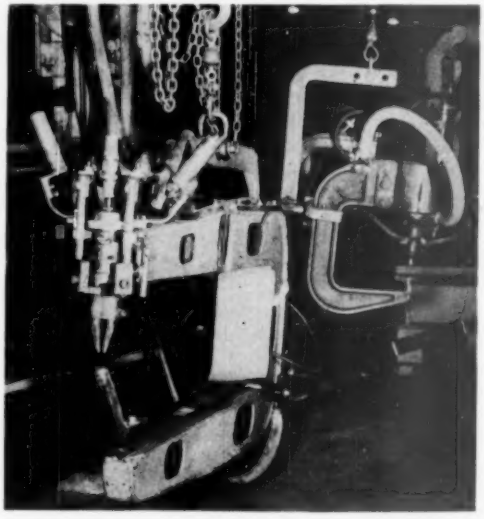


Fig. 7 Double-tip welding tool used to make two welds simultaneously

Fig. 8 Large C-type tool used to weld the side frame to the floor structure on stainless-steel railway cars

problem involved in hanging the completed tool. No matter how light the tool may be, a poor method of hanging can completely nullify the best design. Methods of hanging and welding-cable connection vary from job to job and depend on the clearances and welding conditions existing in each case. For this reason no one method will suffice for all tools. The problem is further complicated by the welding cables which are generally heavier and sometimes more bulky than the welding tool. In order to achieve the best working setup it is necessary for the welding-tool designer to keep in mind the conditions under which the tool must work and make provisions for proper hanging and cable connections.

PRACTICAL SHORT CUTS

From the foregoing it might be assumed that the design and manufacture of portable welding equipment is a difficult task for the average shop. In order to show the smaller user of portable welding equipment the extreme flexibility that can be obtained at low cost, a discussion of possible short-cuts follows.

Design and Fabrication of Tools

It is not always necessary or even desirable to use expensive castings for jobs of short duration or for an emergency setup. In these cases it is often possible to substitute a yoke, cut from steel plate. For jobs of low production and low-duty cycle a copper bus bar and water cooling may be dispensed with. In this case the welding tip may be mounted either in a taper socket machined in the steel yoke or it may be silver soldered directly to the steel. The choice of the method to be used is indicated by the tip material. For jobs of higher production or higher duty cycle requiring the use of a bus bar, the same procedure may be followed with the addition of a copper bus bar to carry the welding current.

Application

It is not necessary to limit a portable setup to one tool. Many times, especially on comparatively low-production work it is desirable to have a number of tools connected to a single transformer, each tool being individually supported. This allows almost continuous use of the transformer and timing equipment, avoiding the necessity of shut-downs while tools are being changed. We have had as many as five tools connected to one transformer. However, practical limitations usually allow no more than three.

Another use of portable tools that has been found advantageous is that of permanently mounting a number of small tools on a bench or table. All of these tools are connected to one transformer and timer and means are provided so that only one tool can be operated at one time. The advantages of this setup are that a number of individual pedestal or bench-type welders are eliminated along with duplication of timing equipment. It also provides maximum flexibility for the small shop since the tools can be changed at any time to suit the requirements of the job at hand. A modification of this method is also available for the owner of pedestal welders in that a portable tool may be mounted on a stand and connected to the transformer of the welder with copper bars. This allows the use of a second welding setup which may be used in conjunction with the original welder for a second operation. The use of this procedure allows the small shop with limited welding equipment to perform many jobs which would normally require additional expensive equipment.

In conclusion, the advantages gained by the use of portable resistance-welding equipment are as follows:

(*a*) Flexibility.
(*b*) Ability of the small shop to use mass production methods with a relatively small investment.
(*c*) Time saving achieved by eliminating the need for moving large assemblies to a stationary machine.

Since the author is associated with a company specializing in the fabrication of stainless steel, most of the discussion has been associated with that material. However, the same factors are present when other materials are fabricated by the use of portable resistance welding and are therefore applicable to the entire field.

It is hoped that some of the ideas and techniques presented in this discussion will make the use of portable resistance welding more attractive to many manufacturers who have not fully realized all of the advantages gained by use of this process.

Shielded Arc Welding of Aluminum

Typical argon-shielded A.-C. arc-aluminum welding operations and the savings realizable in comparison with oxyhydrogen torch welding

by Fritz Albrecht

INTRODUCTION

FOR many years now the oxyhydrogen torch-welding process has been the standard means of welding aluminum and aluminum alloys. This position is now being seriously challenged by the electric processes, and particularly by inert-gas-shielded a.-c. arc welding, in almost all applications. The exceptions to the above statement are the welding of such assemblies to which no cleaning can be applied because of shape, size or location and the welding of very fine and delicate apparatus for which no electric equipment is yet available. For brevity's sake the two above-mentioned processes hereafter will be called "arc welding" and "gas welding." The former term is becoming ever more familiar as referring to this special application, indicating the amount of publicity received by this process. Much work has been done to prove the superiority of arc welding in many instances* on the basis of physical and metallurgical properties. Economically however, its claim to superiority is not so well established. Judging by the paucity of published data this new process appears to have been accepted on faith alone, without accurate knowledge of the economies involved. Likewise it has been rejected on the basis of fragmentary information.

PURPOSE

The data to follow are presented to offer a complete coverage of the economic aspects of arc welding of aluminum and aluminum alloys, using the high-frequency initiated inert-arc-shielded a.-c. arc.

METHOD OF PRESENTATION

This paper is primarily directed toward that large section of the welding industry, the smaller shop. To this segment the heavy investment and operating costs of arc welding have been appalling. Fearing the burden of large fixed charges it has considered arc welding inapplicable to its needs for economic reasons. Because gas welding is still the common denominator of the aluminum welding shop, this classical process is therefore taken as the basis of comparison. The fact that the conclusions are drawn on the basis of the smaller shop does not render them irrelevant to the larger shop if certain modifications are applied. These are mainly in the field of overhead, wage scale and duplication of cleaning equipment. Also, attention is called to Graph III which shows the danger point in terms of utilization. The larger shop can usually maintain a more efficient loading of the equipment and in addition need employ a relatively smaller amount of auxiliary equipment, such as cleaning tanks. In general any of the above factors, with exception of overhead, favor the larger shop, therefore improving the economy of arc welding beyond the level to be demonstrated in this paper.

In the work to be described the quality of welding was held to a level generally known as "aircraft quality." This implies smooth fillets of low concavity and absence of defects detectable by "soft ray" radiography. A change from this standard to "commercial quality" will influence the economy of the two processes. Indications are that in "commercial" arc welding further savings will appear. This is due to the nature of the arc which permits an increase in deposit rate without an undue increase in difficulty of manipulation. Similar increase in deposit rate by torch will require a large flame with a relatively serious increase in manipulation difficulties.

The selection of the assembly on which this investigation was conducted was based on the following: It was considered desirable to conduct the test in a gage in which the relative welding speeds are representative of the majority of the gages handled by single passes. To exclude extraneous influences to the greatest degree in determining relative welding speeds for the various gages, flat butt-weld samples were welded. For each process and method (gas; argon-shielded arc; helium-shielded arc) one set of samples were fabricated, each set consisting of 4 samples for each of the eight most generally used gages. The welding was timed and the results of this test are shown in Table 1 and Graph I.

Fritz Albrecht is with The Glenn L. Martin Co., Baltimore, Md.

Presented at the Thirtieth Annual Meeting, A.W.S., Cleveland, Ohio, week of Oct. 17, 1949.

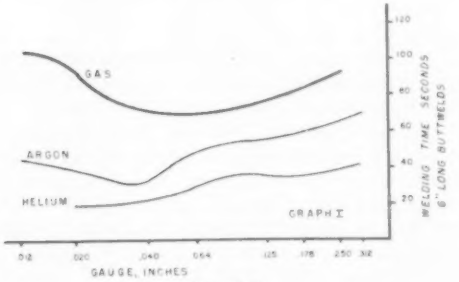
* Voldrich, C. B., "Welded Joints in Thick Aluminum Plates," THE WELDING JOURNAL, **28** (6), Research Suppl., 275-s to 288-s (1949).

Table 1—Welding Time of 6-In. Long Butt Welds in 52 SO Sheet

Time in sec., average of 4 specimens

Gage, in.	Process: Torch welding, sec.	Arc-welding helium arc, sec.	Argon arc sec.
0.012	105	*	45
0.020	92	20	39
0.032	†	†	32‡
0.040	71	24	33
0.064	72	31	50
0.125	77	36	55
0.178	85	37	59
0.250	93	39	65
0.312	*	42	70

* Impractical to run.
† Test omitted.
‡ Run to verify curve, Graph I.



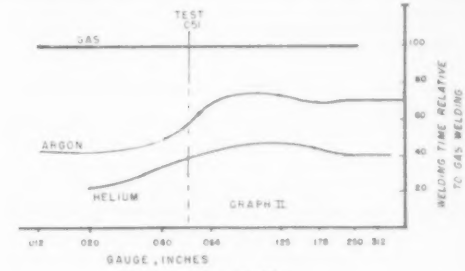
Graph I

Average welding time of 4 specimens vs. gage

The curious "rise" condition which may be interpreted as a result of an inability of the welder to efficiently coordinate manipulation with heat input below gage 0.032 was rechecked. These figures are converted to relative

Table 2—Time Comparison Using Oxyhydrogen Torch-Welding Time of Table 1 As Unity

Gage, in.	Process: Torch welding	Arc welding: Helium arc	Argon arc
0.012	1	...	0.43
0.020	1	0.22	0.43
0.040	1	0.34	0.49
0.064	1	0.43	0.69
0.125	1	0.47	0.72
0.178	1	0.43	0.69
0.250	1	0.42	0.70
0.312	1	0.42	0.70



Graph II

Relative efficiency of arc-welding as compared to gas-welding in terms of welding time

efficiency in comparison with gas welding in Table 2 and Graph II. Gage 0.051 was selected as representative of a mean condition of efficiency. This selection is further justified by the fact that this gage is one of the more frequently used thicknesses. As helium arc welding is not applicable to all types of assemblies due to its bad starting characteristics and high skill requirements, argon arc welding was selected as the method. Furthermore, at the time of this test, helium was not yet com-

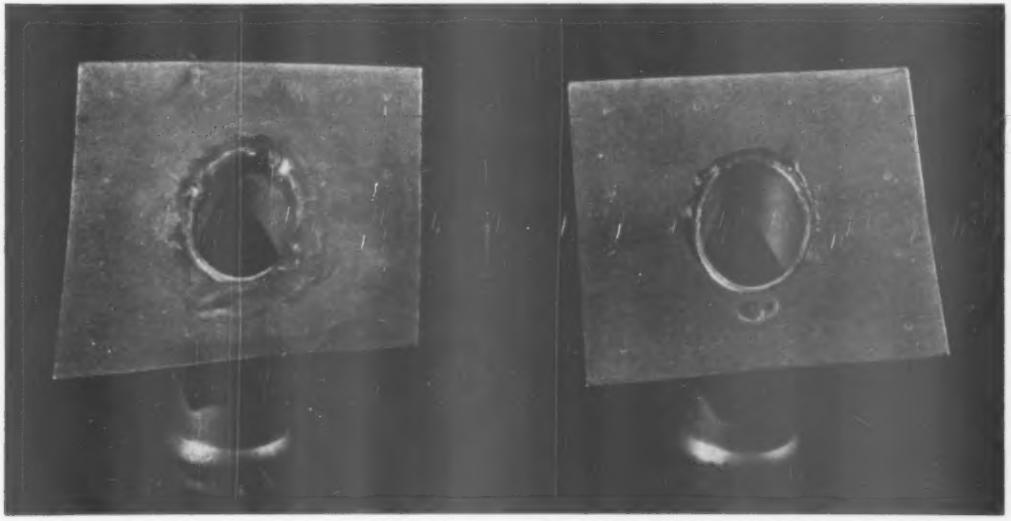


Fig. 1 Production quality weldments, welded with (left) oxyhydrogen torch, showing "burn-through" and warpage heavier than on arc-welded part (right)

Table 3—Time Analysis of Welding 40 Assemblies Shown in Fig. 1 (Time in Hours)*

Operation	Welding process					
	Gas welding, hr.		*Argon arc, hr.*		*Helium arc, hr.*	
Degreasing		0.25		0.25		0.25
Deoxidizing		...		0.53		0.53
Transportation		0.10		0.10		0.10
Flux mixing	0.11		...		...	
Fluxing parts	0.15		...		...	
Fluxing rod	0.25		...		...	
Set-up parts	0.31		0.31		0.31	
Tack weld	0.51		...		...	
Finish weld	1.45	2.78†	0.94	1.25†	0.59	0.90†
Transport to cleaning		0.10		0.10		0.10
Clean Flux and inspect		1.46		...		...
Neutralizing		0.20		...		...
Transport to bench		0.10		...		...
Grind excess and smoothen		1.10		0.35		0.45
Straighten and check contour		1.30		0.75		0.75
		7.39		3.23		2.98

* No helium-welded part shown.
† This represents the welder's time; balance, unskilled, labor.

pletely investigated and not approved for production welding. The fillet type of weld was selected because in actual production fillet welds predominate. If possible, acute and obtuse angle fillet welds should be present on the test piece. All characteristics established above as desirable are fulfilled by the assembly shown in Fig. 1.

Mere welding is rarely the sole determinant of welding economy. Postwelding handling must be considered as a function of, and necessitated by, welding process characteristics. The assembly selected permits scrutiny of this factor. The base, having a double curvature, tests the two welding processes for their warpage producing tendencies. As the warpage has to be corrected after welding by handwork in both cases, its magnitude seriously influences the cost aspect. Similarly the amount of "burn-through" in either process has a decided influence on the economy, rework being demanded to produce a smooth appearance as illustrated in Fig. 2. Jigging in both cases was identical.

The test was run on 40 assemblies welded by gas and argon arc and timed without the knowledge of the operators. This established the time analysis of Table 3. The times so obtained were verified on subsequent lots of the same assembly. Welding time variations so found were less than 10% and the figure used herein is slightly above the average time. Spot checks on other assemblies indicated that the relative speeds found in this 40-piece-lot are comparable to speeds on larger assemblies. This was taken to justify the extension of the 40-piece-lot weld time to 2000 hr.; i.e., one year's effective production time. No actual production check was observed on helium arc welding. The ratio of relative speed in the flat-plate butt-weld test of Table I was applied to the argon weld time to derive a helium arc-weld time. This hypothetical figure was then extended to a year's production for informative purposes only. Spot checks of relatively short runs indicate that no serious discrepancies should be found to exist between helium arc production times and the theoretical projection here presented.

Table 4—Direct Costs of Welding 40 Assemblies

Item	Welding process					
	Gas welding		*Argon arc*		*Helium arc*	
LABOR:						
Welder, $1.60 per hour*	2.78 hr.	$4.45	1.25 hr.	$2.00	0.9 hr.	$1.44
Helper, $1.30 per hour	4.61 hr.	5.99	1.98 hr.	2.57	2.09 hr.	2.72
Total		$10.44		$4.57		$4.16
OVERHEAD, 90% of Labor		9.39		4.12		3.74
GASES*	*Oxyhydrogen:*		*Argon:*		*Helium:*	
	Tacking ..... 0.74 hr. Welding ..... 1.45 hr. No. 3 Tip ..... 2.19 hr. at $0.084 per hour	0.18	9 liters per minute; $0.12 per cu. ft. for 0.94 hr.	2.16	16 liters per minute; $0.085 per cu. ft. for 0.59 hr.	1.78
CLEANING TANK HEAT:*						
300-gal. etch sol. kept at 200° F., 18 cu. ft. gas per hour at $0.004 per cu. ft.		...	For 1.25 hr.	0.09	For 0.9 hr.	0.06
ELECTRIC POWER:*						
$0.049 per kwh.		...	2.4 kwh., 0.94 hr.	0.11	2.4 kw., 0.59 hr.	0.07
WELDING ROD:*						
No. 43 at $0.62 per lb.	2.56 lb.	1.59	1.82 lb.	1.12	2.04 lb.	1.27
		$21.60		$12.17		$11.08

* Assumed industry average.

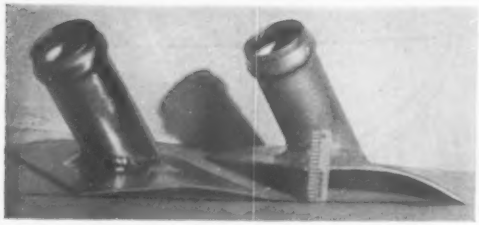


Fig. 2 Production quality weldments welded with (left) arc and (right) with oxyhydrogen torch, showing relative degree of warpage which is negligible on arc-welded part, base plate having slight radius before welding

Table 5 (*A*)—Investment, Interest and Amortization, One Gas-Welding Outfit vs. One Arc Welding Outfit for Either Argon or Helium Shielding

Item	*Investment*	*Life, yr.*	*Cost per year*
	Gas Welding		
Torch	$ 43.55	5	$ 8.71
Regulators	70.00	5	14.00
Welding Tips	3.85	0.5	7.70
Hoses	8.70	2	4.35
Investment	$126.10		
Interest, 4.75% on $126.10			6.00
Fixed cost per year			$ 40.76
	Arc welding		
Holder	$ 85	1	$ 85.00
Power Installation	681	15	46.00
Welding Transformer	1080	5	216.00
Regulator	35	5	7.00
Etching Tank, installed	285	4	71.25
Investment	$2166		
Interest, 4.75% per year on $2166			102.90
Fixed cost per year			$528.15

Table 5 (*B*)—Supplies, Maintenance and Other Charges Not Ascertainable During Test Run, for One Year

Item	*Gas welding*	*Arc welding*
Detergent	$19.65	$32.00
Sodium hydroxide	...	38.00
Water, cooling and rinsing	2.50	4.30
Tungsten electrodes and collets	...	47.00
Nozzles (gas caps)	...	11.25
Maintenance	29.50	19.00
Total yearly supplies	$51.65	$151.55

COMPILATION ON COSTS

Table 3 shows the actual time consumed in welding 40 assemblies by the two processes. Note that certain times are assigned to the welder with the balance expended by nonskilled labor. Table 4 shows Direct Costs which are those directly a function of welding time, which are associated with a particular job and are ascertainable during the timed run. Note the cost of electric power and heating gas. These prices are taken as typical and may have to be adjusted to local conditions. Their influence on the total picture, however, is negligible. The ratio of gas-welding to argon arc welding time in the assembly test is higher by about 7% than in the butt weld test, Table 1. This is within the range of variation found in spot checks and gives a measure of the variations likely to occur. Table 5 (*A*) shows the investment and amortization cost incurred with either method and process during one year. The important aspect of this table is that it represents the "Fixed Charges" which are not related to the amount of production, thus representing an economic burden to be carried by whatever large or small production is achieved during the year. This is in contradistinction to all the other charges, appearing in Table 5 (*B*) and elsewhere. These Direct Costs are essentially a function of production. The above two groups of charges determine the "break-even" point to be discussed later on. Those charges which could not be ascertained during the test run are carried as Supplies and Maintenance. The maintenance time for the gas-welding outfit was obtained with good accuracy by prorating the time of a certain employee who cares for all gas-welding apparatus. The maintenance time for the arc-welding apparatus was obtained by estimate and represents a maximum figure. Nozzle, tungsten and chemical costs were ascertained from usage records.

To convert the 40-assembly-lot time into a yearly cost figure the welder's time in each of the three cases in Table 3 was extended to 2000 hr. The balance of the time in Table 3, the "unskilled labor" portion was expended proportionately. Thus the first item in Table 6, Labor, was obtained. Charges for supervision, shelter, services, etc., are absorbed by Overhead, Item 2 of this table. Direct Costs of Table 4 were similarly expanded

Table 6—Cost of Operation for One Year

Item	*Gas welding*		*Argon arc*		*Helium arc*	
Labor:						
Welder, $1.60 per hour	2000 hr.	$3,200.00	2000 hr.	$3,200.00	2000 hr.	$3,200.00
Helper, $1.30 per hour	3316 hr.	4,311.51	3168 hr.	4,118.00	4644 hr.	6,037.77
Total		$7,511.51		$7,318.00		$9,237.77
Overhead, 90% of Labor		6,760.35		6,586.56		8,313.99
Direct Costs, from Table 4		1,273.38*		5,568.00†		6,666.73‡
Fixed Costs, from Table 5 (*A*)		40.76		528.15		528.15
Supplies, from Table 5 (*B*)		51.65		140.30		151.55§
Cost of one year's work		$15,637.65		$20,152.66		$24,898.19

* Derived from $1.77 for 2.78 hr. expanded to 2000 hr.
† Derived from $3.48 for 1.25 hr. expanded to 2000 hr.
‡ Derived from $3.18 for 0.90 hr. expanded to 2000 hr.
§ Allowance made for larger amount of work processed.

Table 7—Comparison of Welding Costs and Productivity of Three Welding Methods. Basis of Comparison: 40 Pieces As Per Fig. 1 and One Year's Production on This Assembly

	Gas welding	*Argon arc*	*Helium arc*
Time to weld, hr.	2.78	1.25	0.9
Expected production per year, units	28,770	64,000	89,000
Cost of one year's operation	$15,637.65	$20,152.66	$24,898.19
Cost per unit	$0.543	$0.315	$0.28
Fixed Cost per year	$40.67	$528.15	$528.15

as they are a direct function of the welder's time to obtain the third item. The fourth and fifth items, Fixed Costs and Supplies, were initially in terms of one year and are so entered in Table 6. The sum so derived is the total cost of gas and argon arc welding on the basis of direct observation, and the cost of helium arc-welding on the basis of deduction.

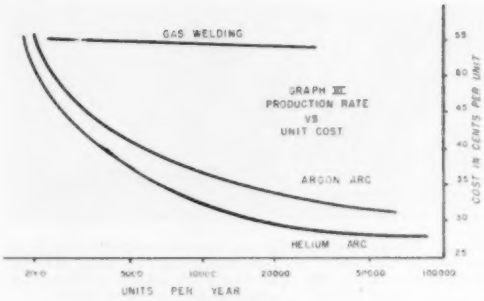
The productivity of the welder is then determined on a yearly basis by expanding the 40-piece-lot to a year's production in terms of the ratio of lot time to 2000 hr. This then furnishes the basis for Table 7 which presents a condensed version of the processes and methods under consideration. While the actual figures of the costs may be modified, the ratio of the costs gives a clear indication of the economies which arc welding affords.

BREAK-EVEN POINT

There is still one point to be settled. This revolves around the fact that in this work it has been tacitly assumed that the equipment is continuously loaded, i.e., utilized to full capacity. This utopian condition only rarely obtains in mass production and certainly never in the type of enterprise toward which this article is slanted, the job shop. Due to the relatively heavy fixed charges of arc welding as compared to the very small fixed charges of the gas equipment, definitely higher loading must be maintained on the arc-welding apparatus to earn the profit to defray the capitalization charges. The question therefore arises: "At what production level do we start to lose money?" The magnitude of safe minimum loading is amenable to mathematical determinations as follows: A solution is sought which gives that number of pieces fabricated by one method at a certain cost and the number of pieces produced at the same cost by another means. The cost being common, an equation may be set up for each of two comparisons, i.e., gas vs. argon arc and gas vs. helium arc. The form of the equation is:

$$Y = a + b/X = a_1 + b_1/X_1$$

where



Graph III

Cost per unit vs. number of units produced in one year

Y = cost.
a = direct cost per unit (Table 7).
b = fixed charges of the method (Table 7).
X = number of units produced.

Solving for X it will be found that at 1800 units yearly production gas-welding and helium arc-welding costs are 56.5 cents per unit, and at 2000 units gas welding and argon arc welding are equal at a price of 56 cents. This is illustrated in Graph III and the basic figures are tabulated in Table 7. Note that the loading must decrease to an extremely low figure before the higher fixed charges of arc welding make this process noncompetitive with gas welding. However, the competitive position of a lightly loaded arc welder is very much poorer than the position of a well-loaded arc welder. This is not the case in gas welding where the competitive position is practically unaffected by loading, due to the low fixed charges.

CONCLUSION

The data presented in the next to last paragraph indicates that arc welding is economically justified without reservation wherever sufficient work is on hand to permit continuous utilization and where a capital expense of about $2200 is possible. The savings are of such magnitude that equipment cost may be paid off in one year's operation.

The last paragraph above indicates that there are definite economical limitations to arc welding if the equipment is not utilized above a certain production level. This level, however, is so low as to make this eventuality quite remote. The employment of helium as the shielding gas will increase the savings of arc welding still further. However, technical problems make helium gas inapplicable in some cases.

Formal Training of Welding Personnel

Advanced technical training is offered for skilled mechanics, technicians and industrial production specialists

by Allen C. Tyler

THE LeTourneau Technical Institute's primary objective is to provide a program of education for skilled mechanics, technicians and industrial production specialists. In our program we have attempted to follow the fundamental steps of psychology of learning more nearly than the average vocational school training program or industrial apprenticeship training program. Our graduates will be capable and mature men in both education and work experience to enable them to be productive, self-supporting, self-directing citizens—happy and contented with the dignity of their vocation.

In the LeTourneau Technical Institute we have the following four-year program for Welders:

		Lecture	Laboratory	Credits
	First Year			
113–123	C. Alg. & Trig.	3	..	6
113–123	English	3	..	6
112–122	Gen. Engr. Draw.	1	3	4
113	Weld. Science	3	..	3
113–123	Economics	3	..	6
111–121	Weld. Lab.	..	3	2
114	Chemistry	3	4	4
				31
	Second year			
213–223	Govt. St. & F.	3	..	6
213–223	English	3	..	6
213	Weld. Science	3	..	3
211–221	Weld. Lab.	..	3	2
124	Chemistry	3	4	4
111–221	Gen. Eng. Draw.	1	3	4
5	Sem. hr. elective	..	..	5
				30
	Third year			
114–124	Physics	3	2	8
313	Fer. Metallurgy	3	..	3
313	Weld. Science (Engr).	3	..	3
311–321	Weld. Lab.	..	3	2
313	Time Study	3	..	3
312	Safety Engr.	2	..	2
313	Quality Control	3	..	3
6	Sem. hr. elective	..	..	6
				30
	Fourth year			
413	Ind. Rel.	3	..	3
423	Ind. Mgmt.	3	..	3
413–423	Prod. Methods	3	..	6
413	Weld. S. Design	3	..	3
411–421	Weld. Lab.	..	3	2
413–423	Tool & Jig	3	..	6
212	Slide Rule	..	..	2
5	Sem. hr. elective	..	..	5
				30

(Welding Mechanic: First 27 mo.;
Welding Engineer: Full 36 mo.):

Electives—Years 2, 3, 4—to make 120 hr.

		Lecture	Laboratory
214	Analytics	4	..
223	Calculus	3	..
214–223	Chem. Quan.	3	4
312–322	Draw. Str. Layout Design	1	4
313	Analysis of Mech.	3	4
314	Ind. Chemistry	4	..
315	Prin. of Machining	4	3
323	Metallography	3	..
323	Str. of Materials	3	..
112–212–312	Bible	2	..
115–125	Weld. Shop Process	..	25
215–225	Weld. Shop Process	..	25

Course Objective: **Welding Mechanic and Welding Engineer**

Expenses—long term (9 mo.)

Not to exceed the following:

	Year			
	1st	2nd	3rd	4th
Fees	$ 35	$ 35	$ 35	$ 35
Tuition	270	270	270	270
Tools	50	35	..	..
Textbooks	45	45	45	45
Equipment	25	15	10	10
Supplies	40	40	40	40
	$465	$440	$400	$400

A four-year program of training is offered such that the graduates will have a broad educational base.

Across the street from the Institute the plant of R. G. LeTourneau, Inc., offers part-time employment up to 25 hr. a week in the trade objective of the student. The work experience is of a varied type, as the plant of R. G. LeTourneau, Inc., is engaged in the production of heavy machinery for earth moving and house building. In the shop the students are under the direct

Allen C. Tyler is Acting Dean, LeTourneau Technical Institute, Longview, Tex.

Presented at the Thirtieth Annual Meeting, A.W.S., Cleveland, Ohio, week of Oct. 17, 1949.

supervision of an instructor. The shop Foreman assigns work and the Institute Instructor sees that the student is properly instructed and able to accomplish the assigned job. The student is paid for the work he does on an hourly rate. The rate of pay and his progress is established by the R. G. LeTourneau, Inc., plant in cooperation with the Institute rating scale. The scale has been approved by the U. S. Wage Stabilization Board. The student has a pay raise every 90 days if his school work, as well as his plant work, is satisfactory. This system of pay rate brings him to a slightly lower rate than he will receive upon graduation.

The student is transferred from job to job and from department to department as long as the work fits his trade objective.

The Institute classwork is of a college grade. One-hundred twenty semester hours of work are necessary for a Bachelor of Science Degree.

Thus the main difference between our program and other school programs is that actual shop experience provides what is normally laboratory experience. First, on production machinery with facilities and equipment far exceeding that possible in vocational school laboratories. The facilities of work time provides an almost ideal learning situation for the student. Second, the laboratory work is an actual employment experience in which the student's labor contributes to the production program. In the average vocational program the laboratory is a preplanned exercise which is confined by limited scope and "make-work" which reduces the incentive required for most effective learning. Third, the wage the student receives aids him financially and the part-time work gives him actual experience with Industry which he may use as a credential when he seeks employment.

The classroom work done during the other half day includes English, Math, Mechanical Drawing, Metallurgy, Welding Science, Economics, Government, Physics and other subjects applicable to the work he does in the plant.

Again let me state that these courses are on a college basis and we are fully recognized as a Junior College and we are putting the last two years of our college work on an accreditable basis which will meet the standards required by accrediting agencies.

The LeTourneau Technical Institute is a nonprofit organization dedicated to an educational objective. The physical plant, classrooms, laboratory, library, Administrative offices, gymnasium, swimming pool, athletic grounds and dormitories are the facilities of the Harmon General Hospital that have been modified for the use of the Institute.

Besides training in Welding, Machine Science, Electrical Maintenance, Mechanical Maintenance, Building Trades, Heat Treating and Metallurgy, Linotype and Lithography are course objectives that we have set up at the present time.

Employment in the LeTourneau Co. plants is not an exclusive objective. The subject matter and course contents are broad enough to prepare the student for any employment opportunity which he may decide to accept in his chosen field. Our course is strongly utilitarian, namely, to make the recipient of the program capable of competing for well-paid employment.

The cost of this type of education is necessarily high as those qualified for staff instructors must have industrial trade experience as well as the usual educational background.

In addition to all this, a positive effort is made to aid in the development of the spiritual and religious life of the student. The founders of our Institute, Mr. and Mrs. R. G. LeTourneau, have a deep conviction that moral, religious and spiritual maturity have a propaedeutic value for good citizenship and good workmanship. The chapel periods and shop meetings on the campus are conducted by trained and capable chaplains and religious advisors.

Thus our objective is not that of a narrowly trained operator but of a broadly trained mechanic or technician. He must have sufficient generalized training so that he can qualify as a well-rounded generally trained craftsman, whether he be in a specialized department of a large factory or a small generalized welding shop where he has to meet all types of welding problems.

The soundness and high standards of the Technical Institute program has recently won its recognition by the Texas State Department of Education and the Association of Texas Colleges as an accredited Junior College. Complete accreditation of the Senior College depends on additional departments and continued growth of the program in the following years.

This program is the closest correlation of school and industry on this level we know of, and the nature of the program should insure success.

Continuation of Study of Welding Engineering

The need of additional specialized training such as welding engineering after graduation has been recognized by the individual, industry and the universities

by Frank W. Scott

YOUNG men, graduating from college with engineering degrees, find that the demands of industry are much more diversified than they had expected. Often, little attention is given to their major field or specialization once they are "on the job." They are Engineers and as such are called upon to answer many questions and to do different types of work. They are capable of doing these things because their college training was very broad and they have a little knowledge of many things, with a sound background of basic principles.

As time goes on, the employer expects the engineering graduate to learn more specifically the demands of his job and be prepared to advance. The young engineer may find that this practical application of his college training emphasizes a minor course of study that he had while in school. At times, a man may find that there was nothing in his college training to help him in the job to which he has been assigned. He must dig out the information for himself from books and magazines.

This demand may not come just a few years after graduation—it may be 5 to 10 yr. later. Not only has the man forgotten some of his college teachings, but many technological advances have been made meanwhile. Future promotions will depend upon the engineer's ability to prepare for the new job.

The Welding Courses at many Engineering Colleges and Universities are principally shop work. Only a few of these offer more than that. However, the application of welding in all fields of engineering is growing, and Mechanical, Electrical, Chemical or Civil Engineers have need of better and more thorough instruction. This is evident to all of us engaged in the welding industry. The use of welding in many industries has been and is being held back by the lack of information and training of men in responsible positions.

This need of additional training has been recognized by the individual, industry and the Universities. The role of the Welding Society is important in helping promote welding and to advance the art.

Frank W. Scott is General Manager, Rod Plant, Marquette Mfg. Co., Inc., Minneapolis, Minn.

Presented at the Thirtieth Annual Meeting, A.W.S., Cleveland, Ohio, week of Oct. 17, 1949.

Members of the Northwest Section of the AMERICAN WELDING SOCIETY are fortunate to be situated in the vicinity of the University of Minnesota, as, with great Universities, the faculty and facilities are available to all groups that seek aid on worth-while projects. The combined efforts of the A.W.S. Educational Committee and members of the faculty of this University have made available to A.W.S. members, University students and others, continuing education in welding and welding processes through annual Welding Institutes.

These Institutes for continuation study in welding are not to be confused with the usual types of "Welding Conferences" that are held at other Universities and colleges or conventions throughout the country. The registrants are nearly all professional or technical persons who are interested not only in refresher courses comprising material that they had presented to them during their professional preparation and training, but also in the advances, discoveries and new techniques. By placing emphasis upon the educational background of those invited to attend these Institutes, we help to keep the well-trained professional and semiprofessional people in step with progress being currently made in their respective fields. It is generally accepted, as a fact, that the scientists and research workers in their laboratories and experiment stations have accomplishments far in advance of practice. It is an important part of our program to bring together the practitioner and the research worker and the inventor. Thus, through lecturers and pictures we benefit from the newest products of the best brains without the usual time lag.

Also, these Institutes are not to be confused with the conventional University Extension work. The Welding Institutes are primarily graduate, while Extension work is primarily undergraduate. The work at the Institute is highly intensive. The University offers a room-and-board proposition, and the course is continuous from morning to night, with classes, seminars and lectures. The adult students are withdrawn from their usual work and environment, and with this intensive application, much material can be covered. This is kept up for 3 days. Ordinary extension classes meet one night a week for perhaps 2 hr. during a quarter or semester. Otherwise, the students are at their ordinary occupations. University extension work ap-

peals to the masses, while continuation study appeals to a selected group of people who are training for leadership in the vocations by which they earn their livelihood. The technical and professional people realize that repeated and continuing training is essential to that leadership.

The choice of the faculty to disseminate the most recent knowledge available is a painstaking task. The leadership is assumed by members of the University faculty. However, these are always supplemented by scholars and experts who are brought in from outside the campus. Often these latter are not college teachers, but outstanding authorities in some specialized fields of welding. The selection of those invited to lecture is usually that of the A.W.S. Education Committee. The actual extending of the invitation to act as one of the faculty is by the University. We have been fortunate in having a generous acceptance of these invitations to serve.

Many have commented upon the number of men that make up the faculty for these Institutes. To the casual observer, the number of faculty members seems entirely out of proportion to the number of students. The explanation lies in the character of the teaching. It is self-evident that no 1, 2 or 3 teachers can be expertly qualified to function in all the highly specialized phases of welding. Therefore, for each such highly technical segment of the whole field, there is brought in an expert in that particular matter. This expert is known to be such by members of the profession. People attending these Institutes are keenly aware of the expert guidance given by these leaders.

Many registrants who attend continuation study Institutes are those who have been given responsible positions, or those who we often say have "arrived." The largest age group is composed of persons between the ages of 40 and 60 (42%). People between 30 and 40 yr. come next (35%). They realize the value of continuing study to maintain their position of leadership.

Announcements of the Institute are sent to members of A.W.S., to the companies that are sustaining members of A.W.S. and other corporations primarily interested in welding. Railroad shops, highway departments, mining companies and steel companies are also invited to send their key men. University students and faculty are invited. Due to limited facilities, registrants must be limited. However, some companies send more than one man to these meetings, paying their expenses and allowing them the time for study.

The costs of these courses of study are nominal—registration, $2.00; tuition, $8.00; total $10, payable in advance. Registration is personal, nontransferable and for the entire period of the Institute. These low costs make the courses within the reach of all interested parties. Out-of-town registrants may room and board on the campus at a nominal cost. This is convenient and pleasant during their stay on the campus.

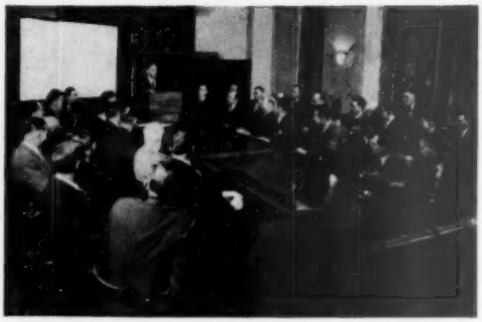


Fig. 1 Typical group of men attending meeting in Center for Continuation Study

The organization of a program that will be worth while to the faculty, engineers, designers and fabricators is an unusually difficult task. The intensity of the continuation study, and the rapid-fire type of presentation make the selection of subject matter of tremendous importance. A "refresher" course, or basic introductory material is a must on the program. In this way, the listener has a better understanding of the advances, or discoveries, that will be presented. The time spent on these "refresher" courses must be limited so as not to use an excessive proportion of the total. It has been found helpful for the planning committee to outline or suggest to the speaker the area he should cover. Through studious care, vague outlines have been made sharp and the discussion given the proper perspective for the subject matter to follow.

After basic principles, the main objectives are reached, that is, what is new in the field under discussion. There is no trouble finding material that might be presented; the difficulty lies in preparing a program that will be most helpful to the people in this section of the country. Many subjects that are presented at annual A.W.S. meetings would have no appeal or little value to the men in Minneapolis and vicinity. The lectures must be applicable to current problems being encountered in production or design. Although new methods, procedures and techniques are subject to the most discussion, fundamentals are emphasized.

Two typical programs are shown (Table I) with the subjects and speakers. Figure 1 shows a group listening to a speaker during a session at the Center for Continuation Study. The informality of these meetings is indicated by this group picture. The fact that the whole group has luncheon and often dinner together breaks down a certain amount of reserve often found among men. This informality and personal contact makes for freer discussion, suggestions and criticism. No formal notes are made at these meetings, promoting open discussion.

The results of these Continuation Study Institutes indicate that they are needed and welcomed by the participants. Requests for further meetings always precede the announcement of coming sessions. Without doubt they advance the science and art of welding.

Table I

UNIVERSITY OF MINNESOTA
Center for Continuation Study

INSTITUTE IN WELDING — OCTOBER 14th, 15th, 16th

PROGRAM

Monday, October 14th

8:00– 9:00	Complete registration. Dormitory Desk, Center for Continuation Study
9:00– 9:30	Introduction, Norman W. Johnson, R. C. Jordan, *Chairman*, Ralph L. Dowdell
9:30–12:00	"Fundamentals of Metallurgy As Applied to Welding," S. L. Hoyt
12:00 M	Luncheon. Center dining room *Chairman*, Melvin Ovestrud
1:30– 2:45	"The Welding of Aluminum and Its Alloys," G. O. Hoglund
2:45– 4:00	"The Welding of Magnesium Alloys," W. S. Loose
4:00– 5:30	"Photoelastic Method of Stress Analysis," H. K. Palmer and O. M. Larsen
6:00 P.M.	Dinner meeting. Center dining room *Chairman*, R. H. Newton *Speaker*, R. E. McFarland

Tuesday, October 15th

Chairman, Chester A. Hughes

9:00–12:00	"Fundamentals of Structural Welding Design," A. T. Waidelich and M. B. Wathen
12:00 M	Luncheon. Center dining room. *Chairman*, R. E. McFarland
1:30– 3:00	"Fundamentals of Resistance Welding," F. A. Hurcomb
3:00– 4:30	"Fundamental Principles and Application of Gas Welding and Cutting," Frank A. McGuire

Wednesday, October 16th

Chairman, C. M. Underwood

9:00–12:00	"Fundamentals of Welding Design Applied to Mechanical Equipment," Robert Davies
12:00 M	Luncheon. Center dining room *Chairman*, Thomas P. Hughes
1:30 P.M.	General Summary and Round Table

FACULTY

ROBERT DAVIES, Consulting Engineer, Lincoln Electric Co. Cleveland
RALPH L. DOWDELL, Professor of Metallography, Univ. of Minn.
G. O. HOGLUND, Jobbing Division, Aluminum Company of America, New Kensington, Pa.
S. L. HOYT, Technical Adviser, Battelle Memorial Inst., Columbus, Ohio
CHESTER A. HUGHES, Associate Professor of Structural Engineering, Univ. of Minn.
THOMAS P. HUGHES, Assistant Professor of Mechanical Engineering, Univ. of Minn.
F. A. HURCOMB, Sciaky Bros., Chicago
NORMAN W. JOHNSON, Head, Center for Continuation Study
RICHARD C. JORDAN, Professor and Assistant Head, Department of Mechanical Engineering, Univ. of Minn.
OTIS M. LARSEN, Assistant Professor of Mechanical Engineering, Univ. of Minn.
W. S. LOOSE, Dow Chemical Co., Midland, Mich.
R. E. MCFARLAND, Western Electric Co., Chicago; Vice-President, AMERICAN WELDING SOCIETY
FRANK A. MCGUIRE, The Linde Air Products Co., Chicago
R. H. NEWTON, *Chairman*, Northwest Section, AMERICAN WELDING SOCIETY
J. M. NOLTE, Dean of University Extension, Univ. of Minn.
MELVIN OVESTRUD, Vice-President and Works Manager, Pioneer Engrg. Works, Minneapolis
HERALD K. PALMER, Assistant Professor of Mechanical Engineering, Univ. of Minn.
C. M. UNDERWOOD, Manager, Northern Ordnance Co., Minneapolis
A. T. WAIDELICH, Assistant Director of Research, The Austin Co., Cleveland
M. B. WATHEN, The Austin Co., Cleveland

UNIVERSITY OF MINNESOTA
Center for Continuation Study

INSTITUTE IN WELDING — NOVEMBER 24th, 25th, 26th

PROGRAM

Monday, November 24th

8:30 A.M.	Complete registration. Dormitory desk, Center for Continuation Study
9:00 A.M.	Introduction, J. M. Nolte, E. E. Hall
10:00 A.M.	"Basic Considerations of Steel Analysis and Weldability," Ralph L. Dowdell
12:00 M.	Luncheon. Center dining room
1:30 P.M.	"Fundamental Considerations When Welding Complex Alloy Steels" (High Sulphur, Air Hardening, Etc.), F. W. Scott
2:45 P.M.	"Newer Techniques In Welding Heavy Gray Iron Castings," Louis J. Larson
4:00 P.M.	"Hard Surfacing Metals As Applied by Welding," Henry S. Jerabek
6:00 P.M.	Dinner. Center dining room *Presiding*, Elmer Smith "Exposing the Hidden Arc," John R. Morrill

Tuesday, November 25th

9:00 A.M.	"Inert Gas-Shielded Arc-Welding Processes" (Heliarc, Argon), Frank McGuire, Jr.
12:00 M.	Luncheon. Center dining room
1:30 P.M.	"Stress Analysis In Weldments," J. J. Ryan
2:45 P.M.	"Welding Sequence In Relation to Stresses," C. M. Underwood
4:00 P.M.	"Fundamentals of Welding Design Applied to Machine Tools and parts," John Mikulak

Wednesday, November 26th

9:00 A.M.	"Fundamental Principles of Welding Design and Its Relation to Cost," John Mikulak
11:00 A.M.	General Summary and Round Table, Thomas P. Hughes, *Chairman*
12:00 M.	Luncheon. Center dining room
1:30 P.M.	General Summary and Round Table, Thomas P. Hughes, *Chairman*

FACULTY

RALPH L. DOWDELL, Professor and Head, Metallurgy, Univ. of Minn.
E. E. HALL, *Chairman*, Northwest Section, AMERICAN WELDING SOCIETY, Minneapolis
THOMAS P. HUGHES, Assistant Professor of Mechanical Engineering, Univ. of Minn.
HENRY S. JERABEK, Assistant Professor, Metallography, Univ. of Minn.
LOUIS J. LARSON, Welding Engineer, Allis-Chalmers Mfg. Co., Milwaukee
FRANK MCGUIRE, JR., Division Engineer of The Linde Air Products Co., Chicago
JOHN MIKULAK, Chief Welding Engineer, American Car and Foundry Co., Berwick, Pa.
JOHN R. MORRILL, Assistant to the Vice-President, The Lincoln Electric Co., Cleveland
J. M. NOLTE, Dean of University Extension, Univ. of Minn.
J. J. RYAN, Associate Professor of Mechanical Engineering, Univ. of Minn.
FRANK W. SCOTT, General Manager of the Rod Plant, Marquette Mfg. Co., Minneapolis
ELMER H. SMITH, President, Smith Welding Corp., Minneapolis
C. M. UNDERWOOD, Manager, Welding Division, Northern Ordnance, Minneapolis

ACKNOWLEDGMENT

The author wishes to acknowledge the cooperation and help of Norman Johnson, Director, Center for Continuation Study, University of Minnesota.

Hot Hardness of Hard-Facing Alloys

Hot hardness has been surveyed at temperatures from 600 to 1200° F. for a number of hard-facing alloys. Factors influencing hot hardness are studied and reported

Howard S. Avery

Abstract

Hot hardness, which is of interest chiefly because of the implied hot wear resistance, has been surveyed at 600, 800, 1000 and 1200° F. for a number of hard-facing alloys, using a Rockwell diamond indentor technique. Some indication of creep resistance, which becomes important as a variable at the higher temperatures, is obtained by incorporating a time factor in the test.

Arc welds, gas welds and castings of wear-resistant alloys are included in the survey, which includes most of the types in the current A.W.S. classification of hard-surfacing alloys. For a few materials creep and stress-rupture data are correlated with hot hardness.

Tempering response, secondary hardening, intrinsic solid solution hot strength and the contribution of hard compounds (chiefly carbides) all influence hot hardness. Low-alloy steels with little resistance to tempering lose hardness rapidly. Alloys such as chromium, tungsten and molybdenum, which cause marked secondary hardening, contribute to good hardness retention up through 1000° F. At 1200° F., solid solution hot strength and primary carbides appear of major importance. The Cr-Co-W alloys generally exceed the other alloys in hardness at 1200° F., but their properties vary widely with carbon content. Since toughness, as indicated by compression tests, varies inversely with carbon content, a compromise between hardness and toughness is required.

INTRODUCTION

HOT hardness is of interest to those who use hard facing to protect industrial equipment against wear primarily because of the implied hot wear resistance. The purpose of this paper is to assemble information helpful in the selection of wear-resistant alloys for elevated temperature service.

The data are presented with a realization that even at ordinary temperatures hardness and wear resistance do not necessarily have a direct relationship. At high temperatures, hardness has even more the status of an unvalidated wear test. Nevertheless, hot hardness information is useful and, until validated hot wear tests are developed, it is necessarily the first recourse for quantitatively ranking different alloys.

Wear is usually a mechanical phenomenon whereby parts of a surface are removed by stresses great enough to cause localized failure of the material. Since strength is the inherent factor opposing such failure, it is reasonable to assume a correlation between wear resistance and strength, or with another property, such as hardness, that is indicative of strength. For relatively homogeneous alloys such as hardened steels the relation of tensile strength to hardness is good, usually permitting estimation of ultimate strength from hardness data with errors under perhaps 10–15%.

Heterogeneous alloys, such as cast irons, do not show an equally dependable relationship. The stresses actually operating on a microscopic scale against a constituent particle or an area beside a discontinuity (such as graphite flakes in cast iron) may be much higher than the nominal applied stress. Marked departures from behavior tentatively predicted by hardness data are frequent when such structures undergo wear testing.

Chemical removal of surface layers is also possible. This is encountered in hot wear applications as hot gas corrosion or scaling. Obviously there is little if any relation between this and hardness. Where the scale is tight and adheres well it may contribute effective wear resistance because of its own hardness. Where it flakes and is easily lost, this weight loss is additive to that from stress wear or abrasion. Both mechanical wear and corrosion can be expressed as weight loss per unit area per unit of time. Of course where selective attack occurs, such as mechanical gouging or intergranular corrosion, surface metal loss units are poor criteria.

Scaling may even complicate the frictional wear pattern since most oxides formed at high temperatures are themselves hard and abrasive. Oxidation may thus provide an abrasive between rubbing faces. Fretting is a type of wear that is accentuated by this mechanism.

Hardness will thus illuminate only one aspect of wear. How effectively it is used will depend on how closely observation can be carried to the microscopic scale where wear actually occurs and on how well judgment is used in assaying other contributing factors.

Howard S. Avery is Research Metallurgist, American Brake Shoe Co. Mahwah, N. J.

Paper presented at the Thirtieth Annual Meeting, A.W.S., week of Oct. 16, 1949.

HARDNESS MEASUREMENT

Hardness can be defined and measured in different ways.[1] It usually is considered to be an index of resistance to penetration, with measurement in terms of the deformation caused by penetration under a stated load.

Well-known tests in this category are the Brinell (10-mm. diameter steel or tungsten-carbide ball), the Vickers (using a symmetrical diamond pyramid indentor), the Knoop (a microhardness technique with an elongated diamond pyramid indentor) and the Rockwell (sphero-conical diamond penetrator or steel ball indentors). Brinell and Vickers hardness are reported in terms of apparent stress (load per unit area after elastic recovery), while the Rockwell scale indicates depth of penetration for an increment of load (Fig. 1).

The Rockwell instrument was selected for the test program reported here because of its many convenient features. The modification of the equipment for hot tests was patterned after that reported by Bishop and Cohen[2] at the Massachusetts Institute of Technology (Fig. 2). The Rockwell "C" scale is widely used for reporting the hardness of hard-facing deposits. It is superior to Brinell hardness in the range above C60 or 650 Bhn. because deformation of the Brinell ball introduces errors and the small, shallow impressions are quite difficult to measure. Both the Rockwell "C" scale and the Vickers diamond pyramid scale have the advantage of diamond indentors. Rockwell tests are better known in the United States and were selected because of the direct reading dial system, the facility with which readings are rapidly made, the small indentations which make practicable several check determinations on small specimens, and the ease with which the time factor can be included.

The "C" scale has a limitation. It is intended for hard materials and is poorly adapted to alloys softer than C20. Since temperature softens most metals, this becomes serious in hot-hardness testing. To retain the advantages of the "C" scale designation as far as possible, the data have been recorded well into the range below zero by using negative numbers. This has been possible for most tests up through 1000° F., but at 1200° F. the hardness of many alloys is so low that the data were generally read on the "A" scale, using the same indentor but with a 60-kg. load. Since the "C" scale numbers are much more familiar to engineers, conversions have been made for the convenience of those reading this paper. In using them the reader should always remember the limitations of conversion, and use the values only for rough comparisons.

Rockwell "C" values can be converted into approxi-

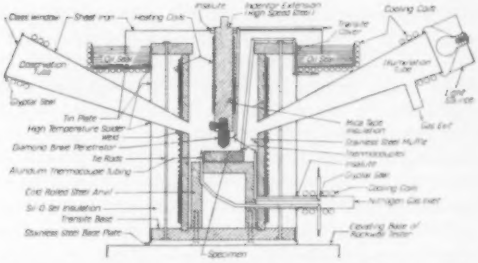


Fig. 2 Cross-sectional sketch of the apparatus used by Bishop and Cohen[2] for Rockwell hot-hardness testing

The same type of equipment was used for this investigation of hard-facing alloys

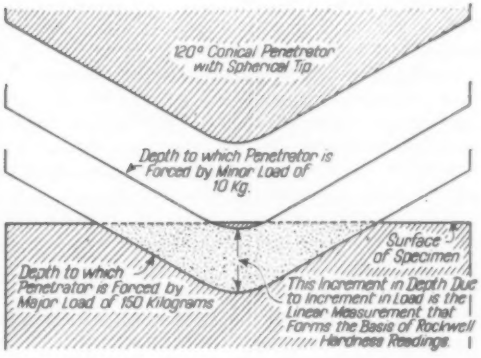

120° Conical Penetrator with Spherical Tip
Depth to which Penetrator is Forced by Minor Load of 10 kg.
Surface of Specimen
Depth to which Penetrator is Forced by Major Load of 150 Kilograms
This Increment in Depth Due to Increment in Load is the Linear Measurement that Forms the Basis of Rockwell Hardness Readings.

Fig. 1 Indentor contour and Rockwell "C" hardness details after Williams[1]

Hardness testing on the Rockwell hardness tester is based upon the measurable residual increment of depth of penetration due to temporary increment of load. In the case of hard steel the penetrator is a diamond cone with spherical tip of minute but definite radius. The first load applied is 10 kg. and the second 150 kg.; so that the increment is 140 kg. The increment of depth due to increment of load remaining after returning to the initial load is 0.00008 in. for each point of hardness on the Rockwell scale. If a piece of steel is as hard as Rockwell C58 at one spot and C55 at another it means that the depth of penetration is three times 0.00008 in. or 0.00024 in. deeper at the softer spot; but that information is given in terms of hardness on the dial

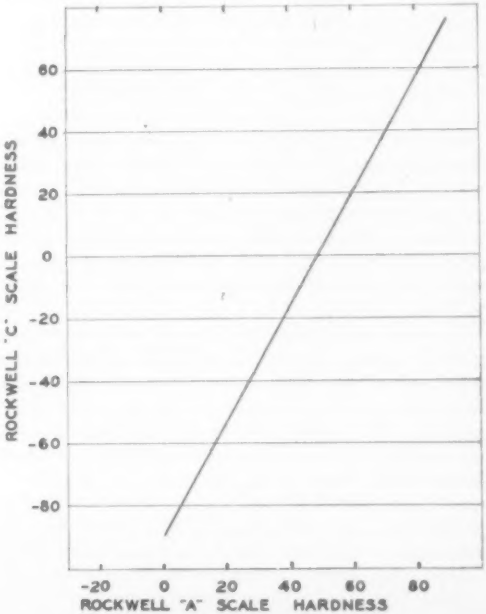


Fig. 3 Hardness Conversion Scale

mate equivalents in other systems, using charts such as Table 1 of A.S.T.M. Standard Designation E48–43T. The Rockwell ranges C20.3–C68.0 and A60.7–A85.6 as they appear in this table have been plotted in Fig. 3 and the line extended to about RA = 0 to facilitate conversion of the low values encountered in hot-hardness determinations. Admittedly, the low values involved lack the precision of the usual Rockwell determinations, especially when deep indentations permit the tested metal to contact the diamond point above the zone where it has been precisely polished. However, the low values are useful for rough comparisons and they highlight the marked softening produced by elevated temperatures.

A number of checks were made to obtain RC and RA values on identical materials selected to provide data in the RA30 to 50 range. These data plot up to five RA points to the left of the line in Fig. 3 and would give converted RC values corresponding higher than those used throughout this report. However, the correlation line has not been altered to conform with these because it is believed that a comparison of this type should be made with an indentor that has been formed and polished further back from the tip than those currently available.

Rockwell hardness testing at room temperature is quite simple. For "C" scale values the suitably prepared specimen is positioned under the conical diamond indentor (Fig. 1), a minor load of 10 kg. is applied to break through scale or surface irregularities, the dial gage is adjusted to zero, and the major load of 140 kg. is smoothly applied through a weight and lever system. The dial gage follows the progress of the penetrator into the specimen, each scale point indicating a penetration of 0.00008 in. The major load is then removed, causing the dial to follow the elastic recovery of the metal. The final dial reading indicates the depth of permanent plastic deformation under the indentor as a result of the stress from the major load. While the final value appears in arbitrary units the test really measures plastic deformation under a fixed load. However, the stress is not constant because the conical shape of the diamond point produces greater projected areas of contact as penetration proceeds. Thus unit stresses fall during penetration until they reach the effective elastic limit (in compression) of the presumably somewhat work-hardened material.

Hardness tests that provide values based on strength may be expected to correlate with more direct strength measurements in compression. Because tensile and compressive properties of steel are similar, the usually available charts relating tensile strength and hardness are justified. However for brittle materials like cast iron and some hard-facing alloys it is not feasible to obtain true tensile strength values; those obtained in conventional tests are probably well below the true values partly because nonaxial loading is almost inevitable. Even slight bending produces surface tensile stresses higher than those indicated by the testing equipment and, because these are not relieved promptly by plastic flow as occurs with steel, fracture is produced by them. Compression tests are not so seriously handicapped by brittleness of the alloy and a better relationship with hardness is expected. Since hard-facing deposits are so frequently stressed in compression, hardness and other compression tests are especially pertinent.

EXPERIMENTAL PROCEDURE

The data herein, when load application is momentary, represent conventional Rockwell tests using the "C" or "A" scales. The Bishop and Cohen[2] heating arrangement illustrated in Fig. 2 was adopted with minor modifications. Essentially the additions to the standard Rockwell tester consist of an electric furnace to heat the specimen, an extension to the indentor shaft to bring the indentor into the hot zone and at the same time keep sensitive parts of the machine cool, a resistance winding to heat the lower part of the indentor holder, a nitrogen atmosphere to protect the specimens and indentor parts from oxidation, ports for observation and manipulation of the specimens, and thermocouple pyrometry.

Actual test temperatures are believed to be within ±10° F. of the reported nominal values. Generally control is held closer than this. Both indentor and specimen, as monitored by individual thermocouples, are within the ±10° F. spread. Since thermal expansion of the indentor shaft assembly is reflected in the sensitive instrument dial gage, temperature variations during the test reading period must be held to the practical minimum or spurious values will be obtained. The combined effect of indentor shaft weight (a constant but unimportant addition to the minor Rockwell load), temperature fluctuations during a test run and variations in specimens of cast material tend to produce a scatter band of about ±1 to 1½ on the "C" scale. This experimental error is reduced by the practice of averaging several readings.

In this program the mounting of the diamond has been a source of trouble. The Massachusetts Institute of Technology workers[2] report no difficulty when using a standard penetrator but their maximum temperature was 1100° F. The practice of testing up to 1200° F. may account for the indentor problems since many materials exhibit a sharp loss in strength between 1100 and 1200° F. A number of indentors were used in the current research. Some involved mounting the diamond in special materials by the manufacturer. Performance was erratic; some gave satisfactory results for many determinations while others failed after only a few tests. Failure occurred by displacement of the diamond in the mount, by cracking of the mount, or by its progressive deterioration. Commercial tank nitrogen was used as the test atmosphere and as this is not always oxygen-free, oxidation can occur with especially vulnerable materials. Some of the special mounts were apparently made of easily oxidized nonferrous metals by powder metallurgy.

Deterioration of the indentor, other than sudden failure, was detected by checking its performance at room temperature against standard hardness test blocks. When the indicated departure from the expected reading exceeded one Rockwell "C" scale point the indentor was replaced. The serious interruptions to a program that can be caused by indentor failures were the primary reason why exploration of hardness was not carried above 1200° F. This is in prospect when indentor improvement renders it feasible.

Since indentor errors are in one direction (for an individual penetrator) they may be additive to those previously noted and they are not minimized by averaged readings. Thus, a possible error of about 2½ "C" scale points is indicated, though most of the reported data are believed likely to be within one scale point of the probable true values, excluding the effects of structural changes, creep or constitutional irregularities in the material under test.

Three different test technique patterns have been employed. The first involved successive tests on one specimen at several temperatures. The specimen was usually tested at room temperature with the unmodified machine and later with the extension indentor assembly, then heated to 600° F., held from 15 min. to an hour to equalize the temperature, and tested. About four or five indentations were made, with records kept of the dial indication at 0-, 1-, 2-, and 3-min. intervals after major load application. After removal of the major load the 3-min. hardness is read directly; the other values were calculated from the difference in dial readings under load. The reported data are averages of these several indentations.

After the 600° F. test was completed, temperature was increased to 800° F., stabilized and the testing repeated. Similar subsequent tests at 1000° F., and sometimes at 1200° F., completed the run. Note that the specimen has been heated to and tested at several temperatures without cooling to room temperature. After the final test, the room temperature hardness is obtained for comparison with the initial value. The difference frequently reflects the tempering that occurred chiefly during perhaps an hour at the maximum temperature, which was usually 1200° F. The specimens were customarily furnace-cooled after the last test.

With this procedure the specimen may begin the run either as-cast, as-welded or after some selected heat treatment. If two conditions are compared the tests are usually made on companion specimens cut from one piece before heat treatment. Occasionally the tested specimen was heat treated, resurfaced and retested.

A second procedure eliminated the successive heating. Individual specimens were heated to the test level, stabilized for a few minutes and tested, after which they were air cooled to room temperature. To survey the effect of four temperatures required four specimens, each of which was then exposed to only one temperature.

The third procedure was planned to minimize or eliminate the effect of tempering during the hot test. The specimens, each to be tested individually at one temperature, are previously stabilized by holding at that temperature for 24 hr. and then air cooled. Thus each piece tested has been previously tempered for 24 hr. at the test temperature. This is perhaps the most desirable of the three patterns and will probably be adopted as standard. A specimen data record, showing how the timed hardness values are obtained appears in Table 1. It is hoped that all the important hard-facing alloys eventually will be surveyed by this technique, though unfortunately not in time for this report.

Table 1—Hot-Hardness Data Record

Test No. 229 Specimen No. R-1850

Material: 2.5% C - 30% Cr - 50% Co - 11% W Arc Weld Deposit
Heat treatment: Stabilized (1200° F.—24 hr.—air cool)
Nominal test temperature: 1200° F.
Start of test—Specimen: 1200° F. Indentor: 1200° F.
End of test—Specimen: 1190° F. Indentor: 1190° F.

Hardness data:

Instantaneous hardness, Rc[a]

39	[a] These readings are obtained by removing the major load immediately after the full major load has been applied.
34.5	
36	
34	
37.5	
Av. 36.2	

Time hardness data	Run No. 1	2	3	4	Av.
Major load applied					
0 min.—Dial reading	−4	−11	−11	−12	...
1 min.—Dial reading	−11	−16	−16	−17	...
2 min.—Dial reading	−14	−19	−18	−20	...
4 min.—Dial reading	−19	−21	−21	−24	...
Major load removed (4 min.—Rc reading)	26	23	23	21	23.3
Dial change (4–2 min.)	5	2	3	4	3.5
Dial change (4–1 min.)	8	5	5	7	6.2
Dial change (4–0 min.)	15	10	10	12	11.8
Calculated Rc readings					
2 min.			(23.3 + 3.5)		26.8
1 min.			(23.3 + 6.2)		29.5
0 min.			(23.3 + 11.8)		35.1

	Time hardness, Rc
0-min. load application	35.1
1-min. load application	29.5
2-min. load application	26.8
4-min. load application	23.3

This hot-hardness survey is part of a broad approach to industrial wear problems. Much related work has been done on castings and advantage has been taken of this to include such data where they are pertinent. Since practically any metal that can be cast can also be made and used as welding rod, the availability of castings from experimental or production heats has expedited the survey and helped to make it more comprehensive.

The castings used were from ½ to 1 in. in thickness and from 4 to 10 lb. in gross weight. The mold cooling rates of such castings approximate those of gas weld deposits on heavy bases or arc welds on such parts when preheating is used. The variables in welding prevent a more precise comparison in terms of cooling curves. Gas welds on small steel bases cool more rapidly and similar arc welds, if no thicker than one layer, cool even faster. The hot hardness of castings, gas welds and arc welds is reported herein, though not always for the same alloy.

The castings have an important advantage of uniformity, making significant the report of a precise chemical analysis. Such data are included with assurance that they are representative. In contrast, weld deposits may be quite variable in composition since carbon and alloys may be burned out during arc welding or carbon may be absorbed during gas welding. Also, with arc welding considerable dilution from the base metal may occur. Since the weld test specimens, many of which were provided through the courtesy of welding rod distributors, usually did not provide enough weld metal for chemical analysis as well as various hardness and wear tests, nominal compositions as obtained from manufacturers' literature or from chemical analysis of rods are generally reported.

Castings were obtained from the Experimental Foundry, from the Electro-Alloys Division, from the American Manganese Steel Division or from the Brake Shoe and Castings Division of the American Brake Shoe Co. Weld deposits were obtained from welding rod distributors or were made under laboratory conditions from experimentally cast rods, from rods purchased on the market, or from rods obtained during production runs of the American Manganese Steel Division welding rod department.

Standardized weld deposits were made on $1^1/_2$- x $2^1/_4$-in. blocks of 1- to $1^1/_4$-in. thick S.A.E. 1020 steel. Current practice is to obtain the wet quartz sand abrasion resistance by wear testing the welded face and then to section the top portion (which includes the overlay) into five $^{11}/_{16}$-in. square by $^{11}/_{16}$-in. thick pieces for hot hardness tests and a sixth from which a cross section through the weld is prepared for microscopic examination.

Sectioning is done with wet abrasive cutoff wheels, selected and operated to produce a minimum alteration of the metal adjacent to the cut. Marked changes in the structure can occur if the specimen is overheated by careless cutting. The small pieces destined for hot hardness tests are finished flat on top and bottom by wet surface grinding or by dressing on an abrasive belt. The final surface finish is usually from a 180 grit abrasive.

The abrasion tests mentioned have been previously described,[2-6] with summaries that include many of the hard-facing alloys for which hot hardness data are now reported. A few abrasion test factors are included here to emphasize certain characteristics of the materials. To interpret these factors the following ranking scale is useful:

Outstanding	Up to 0.20
Excellent	0.20-0.40
Good	0.40-0.60
Fair	0.60-1.00
Poor	Above 1.00

The comparison standard is annealed S.A.E. 1020 steel, which is assigned a factor of 1.00. Numerically, the factors are weight loss ratios. This wear test provides relatively high-stress abrasion and correlated well with ball mill service.[5] Its reliability is such that the factor of one test is within 0.03 of the probable true value in two-thirds of the cases; the average of two tests within 0.02. For a reasonably uniform material, the over-all scatter band is likely to be ±0.05 near the middle of the range of factors generally encountered. Cast austenitic manganese steel, after the toughening heat treatment, over a period of years and in dozens of tests has exhibited a factor range from 0.75 to 0.85.

VARIABLES INVOLVED IN HOT-HARDNESS TESTING

High temperatures introduce mechanical and metallurgical variables. The substitution for elastic behavior of plastic flow (creep) and the effect of temperature and time on the structure of alloys markedly influence the apparent hardness obtained. Thus hot-hardness values cannot have the same status as those from cold tests. The data are quite sensitive to experimental procedure and must be interpreted accordingly.

Ordinary hardness values of ferrous materials generally are stable in one sense, being practically unaffected by duration of loading. The time effect, while detectable, is usually less than one "C" scale point. At 1000° F., this is no longer true. Penetration is not positively arrested by stress falling to the elastic limit because no definite limit can be demonstrated. Even light loads produce continuing plastic flow, which is known as creep. The result is a dependence of the apparent hardness on creep tendencies and upon the amount of creep that is permitted by the test technique.

Past research has not explored this area particularly since the techniques have generally involved a fixed time of load application. Fetz[7] used an instantaneous blow from a drop hammer, which represents "dynamic" testing practice. Bens[8] depended on the short but automatically controlled loading interval of the Vickers instrument. Harder and Grove[9] employed a 30-sec. loading interval to obtain Brinell hardness numbers by the mutual indentation method. These three significant papers each list a number of references that will facilitate the work of those who wish to further explore the handling of this variable in the literature.

Creep is a time-dependent property. When a constant stress is applied, the deformation induced characteristically proceeds at a relatively high but steadily decreasing rate (Stage I), then stabilizes for a period at a minimum rate (Stage II) and, finally (if under tensile stress) accelerates steadily (Stage III) until fracture occurs. Stress of course influences these rates. They may be scarcely measurable under light loads, and Stage I may be so extended that Stage II does not appear during the arbitrary test period.

During a hardness test two factors are influencing creep. The decreasing rates of Stage I are more rapidly decreased by the lowering of unit stress as indentation proceeds. Thus the point where rates become too low to notice is rather quickly obtained. However, if the temperature is high enough to permit creep to occur it will affect the hardness indication in a time-dependent

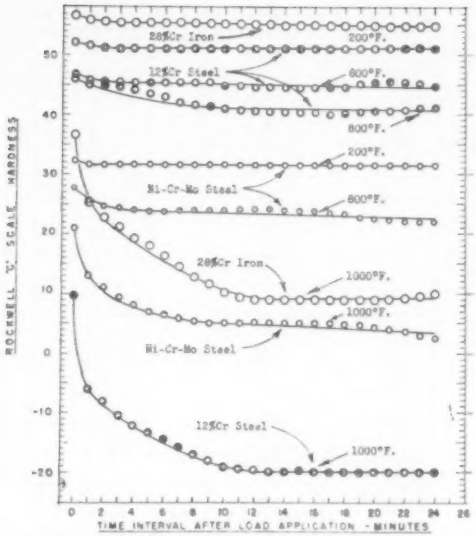


Fig. 4 The effect of time on apparent hot hardness of three iron-base alloys showing creep phenomena at the higher temperatures

pattern illustrated in Fig. 4. Note that the creep occurring in several minutes at 1000° F. can reduce the hardness indication by 20 points. This effect levels off in about 15 min., the changes thereafter being largely masked by the experimental error of measurement. Three or four minutes of load application permits identification and estimation of the creep tendency; therefore, these time intervals are used frequently in this paper.

Tempering is a metallurgical factor that may control hot hardness. Those ferrous alloys that are hardened by austenite transformation, if cooled fast and far enough will attain a maximum hardness that depends on their carbon content (Fig. 5 (*A*)).[10] Reheating this hard structure will soften it to a degree determined by the time and temperature of such tempering. For simple- and low-alloy steels the relationship appears in Fig. 5 (*B*).[11] Most of this effect occurs quite rapidly and perhaps 65% of the softening occurs within 10 min. at 1200° F.[12,13] If hot-hardness determinations are made on a hardened but untempered structure the test temperature is softening it at the same time it is influencing the hardness level and creep. This poses the problem of deciding if initially hard structures should be stabilized by intentional tempering before hot testing. This is certainly advisable for some experiments, as represented by Fig. 4. No uniform practice characterizes past work reported in the literature. To facilitate comparisons both procedures are represented in this report.

With the exception of high-speed steels, for which the conventional secondary hardening treatments around 1000–1100° F. serve to minimize the effect of tempering

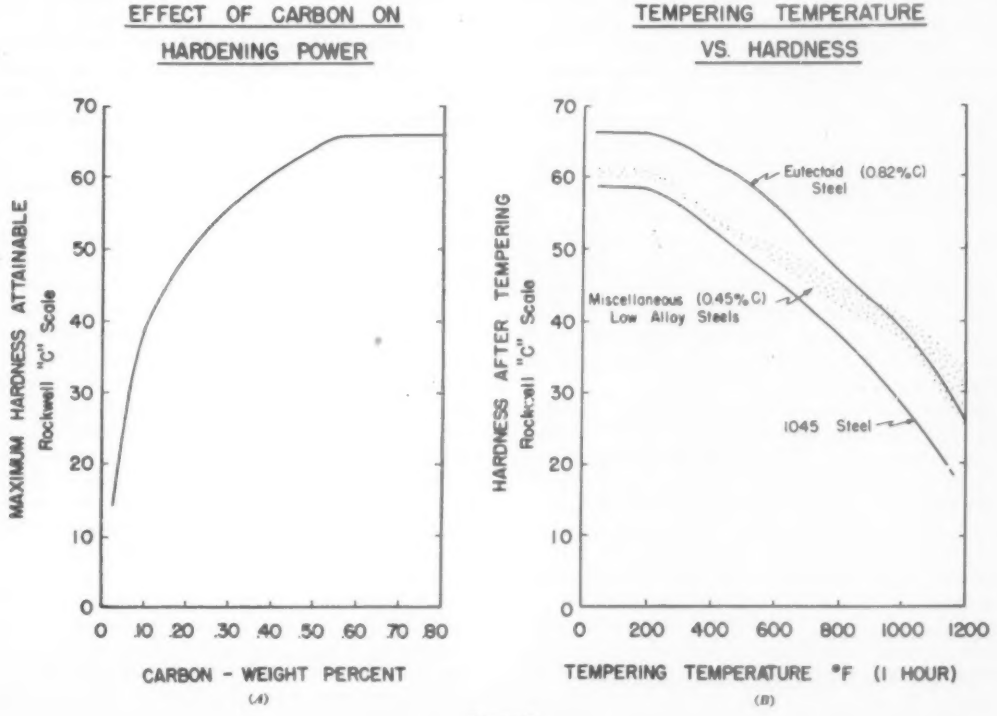


Figure 5

during test runs up through 1000° F., most data available in the literature probably reflect somewhat the structural changes that occur during the hot test. The hardenable ferrous alloys being characteristically affected, this factor should be remembered in making comparisons and evaluations.

Figure 6 shows the pattern that may be encountered. This material is Thermalloy HC-250; a heat- and abrasion-resistant cast iron usually made within the ranges of 2.0–3.0% carbon and 25–30% chromium. Structurally, it contains hard chromium-iron carbides in a steel matrix. After hardening this matrix is martensitic, though usually with some retained austenite. The conventional hardness test, reflecting the combined effect of matrix and large carbides, gives values above C60. From room temperature to 800° F. there is a gradual softening by temperature; reducing the apparent hardness to about C53. At 1000° F. a large drop appears and the effect of creep, which up to 800° F. was minor, becomes pronounced. This test was made on one specimen, heated successively to 600, 800 and 1000° F.

When the survey is repeated on a companion specimen, hardened and also tempered for 24 hr. at 1000° F., the same pattern of behavior is repeated but all of the hardness values are lower. The 1000° F. tempering has reduced the initial hardness from C63.5 to C45.5 and the same drop of about 18 points follows through up to 800° F. By inference, overheating in service or sustained exposure at a constant service temperature may markedly affect the hardness of ferrous materials.

At 1000° F., not only is the lower hardness of the tempered structure apparent but the effect of creep is accentuated. At both 800° and 1000° F., there is some indication that the tempered structures are more vulnerable to creep. The range in apparent hardness at 1000° F. is rather startling. Dependent on the previous thermal history and the time of load application, values from C45 down to C4 can be obtained. If the data were obtained on a hot specimen but with a cold indentor, as is occasionally reported, the range probably would be greater. Thus, unless experimental conditions are precisely reported, hot-hardness numbers may be meaningless or misleading. Even with well-defined techniques the interpretation of the hardness results in terms of wear is difficult.

Service conditions may vary as widely. Ranging from even, sustained temperatures as in some furnaces, to hot cutting and forming operations where a cold tool or part comes in brief contact with hot metal, the requirements of use may be pertinent to the entire spread that hot tests may show.

The high chromium iron of Fig. 6 can be used to describe another factor not yet covered by hot wear and hardness tests but which has been identified in cold abrasion tests. The two major constituents, typed briefly as hard carbides and steel matrix, may change in relative importance. If the carbides can take the brunt of the tendency to wear, acting as hard inserts, the performance may be better than the hardness data might predict. Conversely, where the softer matrix can be undermined, as by an abrasive softer than the carbides but harder than the martensite matrix, the differential wear may cause loss of the carbides by tearing them out or by fracturing them, thus leading to lower wear resistance than would be expected. Since many excellent hard-facing alloys consist of similar heterogeneous structures this complication requires frequent consideration.

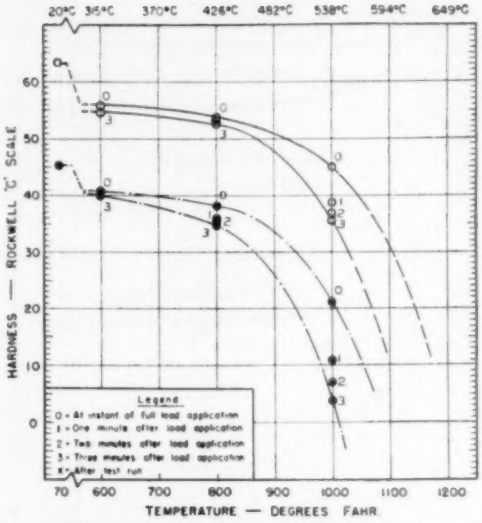
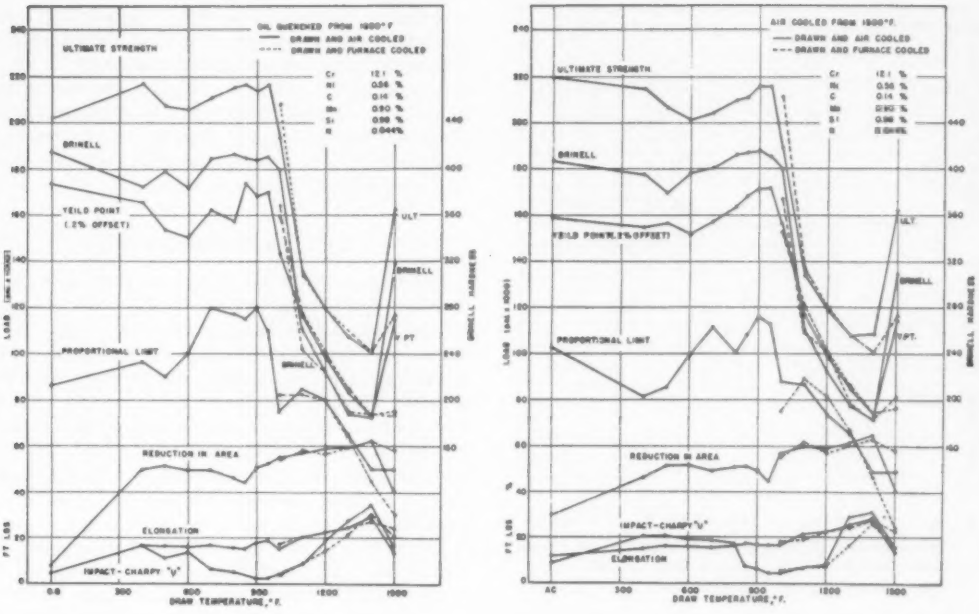

20°C 95°C 370°C 426°C 482°C 538°C 594°C 649°C
HARDNESS — ROCKWELL "C" SCALE
TEMPERATURE — DEGREES FAHR
Legend
0 = At instant of full load application
1 = One minute after load application
2 = Two minutes after load application
3 = Three minutes after load application
K = After test run

Fig. 6 The effect of creep, as controlled by duration of load application, on apparent hot hardness

Hot hardness of a cast 28% Cr iron, after hardening—1950° F.—6 hr.—oil quench—(upper curves) and after tempering—24 hr. at 1000° F.—(lower curves) the material hardened as above

The tempering pattern of Fig. 5 (*B*) is modified by the presence of certain carbide forming elements, notably chromium, molybdenum and tungsten. They induce precipitation of special carbides[14] at temperatures near 1000° F. and superimpose a secondary hardening tendency upon the softening effect of temperature. The result is that for some alloys, notably high-speed steel, heating at 1000 to 1100° F. may actually increase hardness. At 1200° F. the softening tendency again dominates and hardness may be reduced sharply. This pattern is shown in Fig. 7 for a 12% Cr steel.[15] (This material is used as an example since a number of hard-facing alloys use 12% Cr as a base because of the capacity to air harden, as shown by comparing the two charts, and thus to provide hard weld deposits without subsequent heat treatment.) Hot hardness tests may provide evidence of the same effect. Practically, secondary hardening is used to attain hot wear resistance or edge holding ability at the temperatures produced at tool points. This is usually referred to as red hardness. The same metallurgical factor, with its maximum usefulness in the range from 800–1100° F., can be invoked for hard-facing deposits.



(A) (B)

Fig. 7 Properties of cast 12% chromium steel vs. tempering temperature

These charts from Alloy Casting Institute research provide evidence of secondary hardening with a maximum effect near 900° F.

(A) Mechanical properties of base CA alloy oil quenched from 1800° F. then drawn

(B) Mechanical properties of base CA alloy air cooled from 1800° F then drawn

THE EFFECT OF TEMPERATURE

Beside the tempering effect, which can be minimized by pretempering for 24 hr., a characteristic pattern emerges as ferrous alloys are tested at various temperatures. This is exemplified in Figs. 4 and 8. Moderate softening that is temporary and depends on test temperature occurs up to about 800° F. Cooling to room temperature restores hardness to substantially the original value. At the test temperature a minor amount of creep occurs but does not seriously change the hardness as the load is sustained.

Between 800 and 1000° F., or between 1000 and 1200° F. for irons and steels that exhibit strong secondary hardening, there is a sharp drop in hardness and creep begins to dominate the picture. This transition range is important as it generally characterizes iron-base alloys. Tensile tests of steels and cast irons that are not austenitic follow much the same pattern and they thus lose most of their value for predicting load-carrying ability. The behavior in this temperature range is associated with the change from practically elastic to essentially plastic deformation under moderate stress.

Considerable experience in elevated temperature testing outside the field of weld deposits suggests that hot-hardness data will have maximum usefulness at and below 1200° F. Above this temperature time and plastic flow are so important that other procedures such as stress-rupture and creep tests are needed to evaluate the capabilities of an alloy. Certainly they are neces-

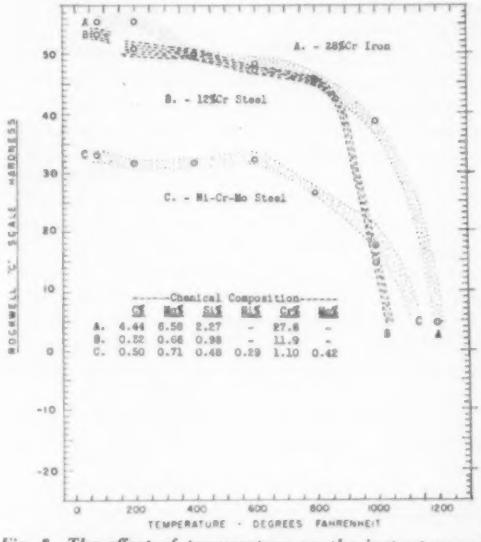

A. - 28%Cr Iron
B. - 12%Cr Steel
C. - Ni-Cr-Mo Steel
ROCKWELL "C" SCALE HARDNESS
TEMPERATURE - DEGREES FAHRENHEIT
Chemical Composition
C% Mn% Si% Ni% Cr% Mo%
A. 4.44 6.58 2.27 - 27.6 -
B. 0.32 0.66 0.98 - 11.9 -
C. 0.50 0.71 0.48 0.29 1.10 0.42

Fig. 8 The effect of temperature on the instantaneous hardness of three iron-base alloys

The specimens were cut from small castings and tempered 24 hr. at the indicated temperatures before testing. The plotted points are averages of five tests

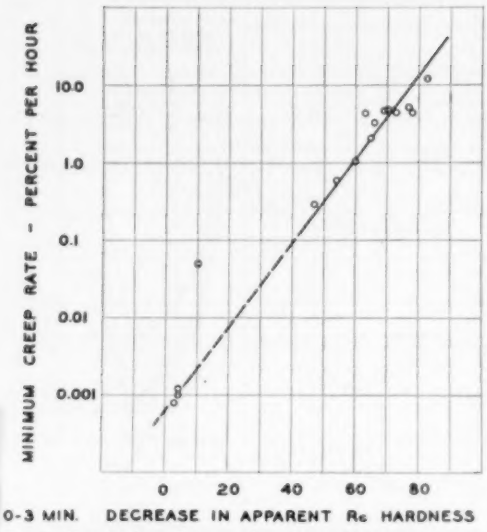


Fig. 9 Correlation of creep data from 20,000 psi. stress rupture and Rockwell hardness tests at 1200° F.

sary to provide design data for sustained load-carrying ability at high temperature. This is not particularly pertinent to hard facing as such conditions would be likely to cause excessive deformation of the base metal before the hard face is affected.

In case the hard facing is applied to a competent heat-resistant alloy the sustained hot strength of the weld becomes important. Only a few hard-facing alloys are suitable for such service, notably the cobalt-base, the nickel-base and the austenitic chromium-nickel-iron types. Their merit is shown by their relative hardness and creep resistance at 1200° F., and has been confirmed by considerable field experience. In addition to creep strength they have good oxidation resistance, which is a property increasingly needed as temperatures exceed 1200° F.

It is inadvisable to generalize about hot hardness in the range from 800 to 1200° F. It is in this zone that ranking may change abruptly with temperature, emphasizing the need for quantitative test data. The tables and charts included here were prepared to permit quick estimation of the behavior of a particular type. To facilitate rigorous comparisons, the data are marked to identify cast, arc-welded and gas-welded specimens, as well as which of the three test techniques was employed.

There will be some interest in the usefulness of hot hardness for predicting creep behavior. Bens[8] included correlation graphs relating fracture time in stress-rupture tests to Vickers hardness at 1600° F. This area is included in the American Brake Shoe Co.'s study of high-temperature factors that cause industrial wear but detailed reporting will probably be the subject of another publication. The chart of Fig. 9, which was prepared by C. R. Wilks, is suggestive of the relations that may be involved. Note that creep rate from conventional stress-strain-rupture tests is plotted against the change in apparent hardness that occurs during 3 min. under load at 1200° F., which is an unconventional use of the Rockwell instrument. It was considered likely that this evidence of creep in the hardness test would be more closely related to tensile creep rates than the actual hardness levels involved.

HOT GAS CORROSION

Scaling in air or the usual industrial combustion gases is seldom a problem below 800° F. The scaling of iron-base alloys is not obviously progressive and the small amount of oxide that forms is not a serious source of metal loss. However, this oxide is hard and may become an effective abrasive under some conditions of wear. Metal to metal friction faces or assemblies that permit fretting may suffer accelerated wear because of oxide abrasion. In extreme cases it may become advisable to minimize oxidation by providing an inert atmosphere.

Above 800° F., scaling is more obvious. The thin films, that from 350–800° F. are usually noticed as temper colors, become a more uniform gray and progressively thicken at higher temperatures. For simple irons and steels, this scale is brittle and can be broken off. The volume changes during formation of the scale may even cause it to crack and flake. Several elements, notably chromium and silicon, change the character of the scale to make it tough, adherent and resistant to mechanical damage. Greater resistance to progressive scaling is associated with these effects. Chromium has thus become a vital component of most heat-resistant alloys.

Nickel also develops protective scales, and while its effect is less potent (perhaps a third that of chromium) it also contributes a valuable austenite stabilizing effect to iron-base alloys. Since austenite is stronger than ferrite at high temperatures, the chromium-nickel-iron group now dominate the field of heat-resisting metals for

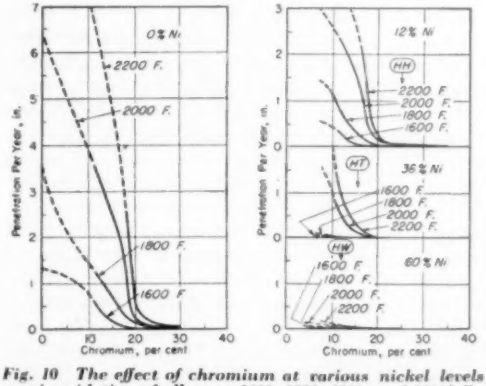


Fig. 10 The effect of chromium at various nickel levels on air oxidation of alloys at 1600, 1800, 2000, and 2200° F., from Alloy Casting Institute sponsored research

industrial service. Their use as welded overlays is a logical outcome of this.

Twelve per cent chromium will provide excellent resistance to progressive scaling in air at 1200° F. and frequently 5% is adequate for practical needs. Above 1200° F., the content of chromium and other protective elements should be proportioned to the exposure temperature. The relation is shown in Fig. 10 for chromium-nickel-iron alloys. For additional details reference should be made to various reports of research sponsored by the Alloy Casting Institute, whose program has included an extensive survey of composition vs. hot gas corrosion resistance.[16–18]

Oxidation rates can be expressed as weight loss (or gain) per unit area or as penetration below the original metal surface. Unless selective attack occurs a penetration rate, as used in Fig. 10, is equivalent to depth of surface metal lost by conversion to scale. In Fig. 10 note that 25% Cr confers a low penetration rate up to 2000° F., and even at 2200° F. the rates are not excessive when 25% Cr is reinforced with nickel. The data indicate that 25% Cr with 0–12% Ni, 20% Cr with 36% Ni and 15% Cr with 60% Ni are desirable minima for service where oxidation resistance is important.

INDIVIDUAL TYPES OF HARD-FACING ALLOYS

During the preparation of a chapter devoted largely to hard surfacing for the Third Edition of the *Welding Handbook*, it became evident to the committee that current opinions about the resistance of hard-facing alloys to hot deformation, while sometimes valid, were not well grounded on quantitative data. In preparing a rating chart the committee used a decile ranking scale to approximately describe various hard-facing materials that were classified through the activities of another committee.

Preparation of the present paper has been partially motivated by the need implied above. To make the test information more useful to those who wish to correlate it with the *Handbook* material, the following discussions of hard-facing alloys have been arranged in the same general order and have been coded with the same reference symbols.*

In a few cases the *Handbook* classification, with its arbitrary composition ranges, tends to put a well-recognized type in two categories. Where this occurs, as in the case of the 5% chromium steels, the types have been described separately.

Simple Carbon Steels: A.W.S. Class I-A-1

The low-alloy content (usually under 1%) makes very rapid cooling necessary for effective hardening of these steels. In weld deposits they are expected to provide ferritic or pearlitic structures of relatively low hardness. They are seldom of interest where hot hardness is important because they temper readily to low-hardness levels, even if initially hardened.

* The coding system appears in Table 1, page 472, and the composition ranges are detailed in Table 11, page 869 of the A.W.S. *Welding Handbook*, 3d edition, (1950).

Table 2—Hot Hardness of Low-Carbon Low-Alloy Casting

A.W.S. Class I-A-2b

Heat No.	*C*	*Mn*	*Si*	*Ni*	*Cr*	*Mo*	*Cu*
	Chemical Analysis, %						
49-004	0.22	0.50	0.50	0.60	0.61	1.02	1.49

Heat treatment 1800° F.—2 hr.—air cooled.
Separate specimens tested at indicated temperatures without previous tempering.
Room temperature hardness: RC 33 before and RC 34 after testing at 1200° F.

Loading interval, min.	*600° F. RC*	*800° F. RC*	*1000° F. (RC)*	*1000° F. RA*	*1200° F. (RC)*	*1200° F. RA*
	Rockwell hardness					
0	32	28	19	59	− 2	48
1	31	26	17	58	−19	39
2	30	25	16	58	−23	37
3	29	24	16	57	−26	35

NOTE (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

Low-Alloy Steels: A.W.S. Class I-A-2

The addition of several per cent of elements such as nickel, chromium, molybdenum, etc., confers some hot hardness, resistance to tempering and the ability to form hard transformation products during normal cooling of a weld. Structures are likely to consist chiefly of very fine pearlite, bainite, martensite or mixtures of these. The group includes several tool and die steel types and the 5% Cr – ½% Mo steels widely used for medium temperature service. Maximum hardness (provided alloys and cooling rate produce martensite) depends on carbon content (see Fig. 5 (*A*)) while the alloy content will determine tempering resistance.

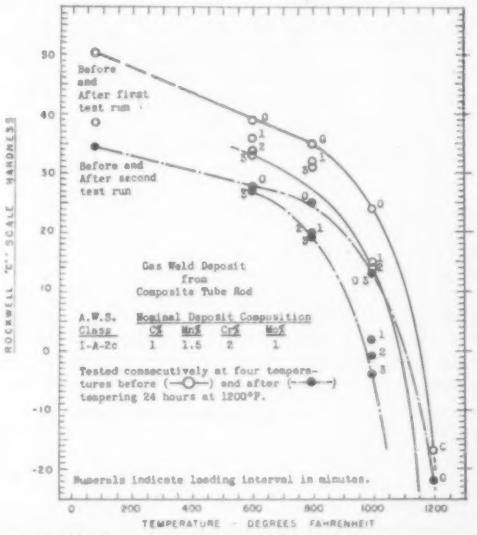

ROCKWELL "C" SCALE HARDNESS
TEMPERATURE - DEGREES FAHRENHEIT
Before and After first test run
Before and After second test run
Gas Weld Deposit from Composite Tube Rod
A.W.S. Class I-A-2c
Nominal Deposit Composition C% 1 Mn% 1.5 Cr% 2 Mo% 1
Tested consecutively at four temperatures before (—○—) and after (—●—) tempering 24 hours at 1200°F.
Numerals indicate loading interval in minutes.

Fig. 11 Hot hardness of a gas weld deposit from a composite tube rod of the "air-hardening" or "self-hardening" type

The upper curves are as welded, while the lower curves show the different properties after tempering 24 hr. at 1200° F.

Table 3—Hot Hardness of Medium-Carbon Low Alloy Casting

A.W.S. Class I-A-2b

Heat No.	Chemical Analysis C%	Mn%	Si%	Ni%	Cr%	Mo%
48–180	0.50	0.71	0.48	0.29	1.10	0.42

Separate specimens tested at indicated temperature after tempering 24 hours at the same temperature.
Room temperature hardness: C32-34 before and C23 after testing at 1200° F.

Load interval in min.	Rockwell Hardness — Testing Temperature 200° F. RC	400° F. RC	600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	31–32	31–33	32–34	26–28	17–21	−28	33.8(a)
1	30.8(a)	30.9(a)	30.9(a)	23.8(a)	10.6(a)	−59	17.3(a)
2	30.6(a)	30.7(a)	30.6(a)	23.4(a)	8.7(a)		11.8(a)
3	30.5(a)	30.5(a)	30.4(a)	23.1(a)	7.6(a)		9.5(a)
4	31.2	31.4	31.2	24.0	8.0		5.5
6	31.2	31.4	31.2	23.7	6.4		2.5
8	31.2	31.5	31.2	24.0	5.2		1.5
12	31.2	32.0	31.9	24.2	5.0		−1.0
16	31.2	32.2	32.0	23.7	5.0		−3.0
20	31.2	32.4	32.0	22.5	4.2		−6.0
24	31.2	32.7	32.0	22.0	2.5		−8.0

(a) indicates the average of four tests.

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

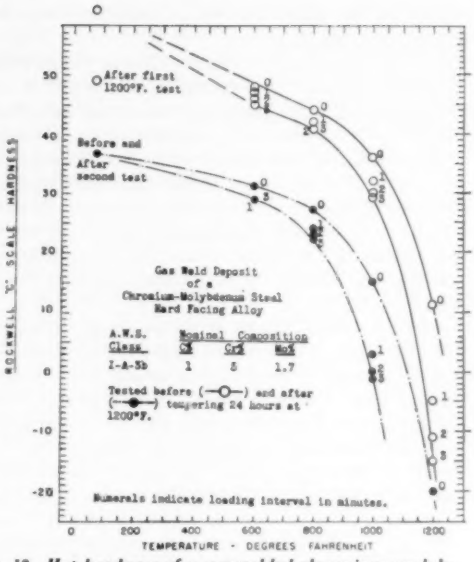

After first 1200°F. test
Before and After second test
Gas Weld Deposit of a Chromium-Molybdenum Steel Hard Facing Alloy
A.W.S. Class
Nominal Composition
I-A-3b
1
5
1.7
Tested before (—○—) and after (---●---) tempering 24 hours at 1200°F.
Numerals indicate loading interval in minutes.
ROCKWELL "C" SCALE HARDNESS
TEMPERATURE - DEGREES FAHRENHEIT

Fig. 12 Hot hardness of a gas welded chromium-molybdenum steel before and after tempering 24 hr. at 1200° F.

To include several carbon levels, a 0.22% carbon Ni-Cr-Mo-Cu steel (Table 2), a 0.50% carbon Ni-Cr-Mo steel (Table 3) and a 1.0% carbon Mn-Cr-Mo type (Fig. 11) were selected to represent this group. The first two are closely related to those compositions frequently employed to pierce billets in making seamless tubes. This service may involve contact with steel at perhaps 2000° F. for a fraction of a minute under high stress, followed by water cooling. Performance is usually satisfactory, though hard facing with another alloy is sometimes tried. Generally hard facing is uneconomical. The compositions are pertinent, however, as they may be used to hard face mild steel where a combination of hot wear resistance and toughness is required.

Figure 11 refers to a welding rod of the class sometimes called "air-hardening" or "self-hardening." The rods carry enough alloying elements to produce a deposit with a hardness usually above 500 Brinell aswelded, though not all of the rods of this class consistently give martensitic structures. The material tested also is a composite tube rod deposit. These rods

Table 4—Hot Hardness of Some 5% Chromium Steels

A.W.S. Classes I-A-2b, I-A-3a, and I-A-3b

Chemical composition, % C	Mn	Si	Ni	Cr	Mo	W	Rockwell "C" scale hardness R. T. (b)	1000° F. 0	1	2	4	R. T. (a)	R. T. (b)	1200° F. 0	1	2	4	R. T. (a)
Small mold cooled castings, tempered 6 hr. at 1200° F.																		
0.21	0.75	1.54	0.55	5.02	0.51		29	8	−6	−9	−12	30	29	(−24	−54	−63	−71)	29
Similar but heat treated: 1800° F.—4 hr.—air cool + 1250° F.—6 hr.—air cool + 1600° F.—4 hr.—air cool + 1250° F.—6 hr.—air cool.																		
0.21	0.75	1.54	0.55	5.02	0.51		28	6	−9	−13	−17	28	28	(−27	−68	−81	−92)	26
Similar to the above with the quadruple heat treatment + 24 hr. stabilization at the test temperature.																		
0.21	0.75	1.54	0.55	5.02	0.51		29	8	−6	−10	−13	29	29	(−24	−61	−69	−78)	29
Castings, heat treated: 1850° F.—2 hr.—air cool + 1250° F.—6 hr.—air cool + 24 hr. stabilization at 1200° F.																		
0.28	0.51	0.44	0.56	4.93	0.51								22	(−30	−74	−88	−100)	21
0.30	0.39	0.45	0.60	4.96	1.60	1.36							26	(−19	−55	−64	−75)	25
Arc weld deposits; stabilized 24 hr. at the test temperature before testing.																		
0.60 (nominal)				6.00	0.5		52	49	31	29	27	51	39	(−9	−45	−56	−68)	38
0.60 (nominal)				6.00	0.5		42	25	21	20	19	41	38	(−11	−33	−39	−47)	32
0.60 (nominal)				6.00	0.5		46	34	32	31	31	47	41	(−5	−26	−32	−41)	34
0.60 (nominal)				6.00	0.5		45	32	27	26	25	43	32	(−16	−45	−54	−61)	32
1.0 (nominal)				5.00	1.7		49	35	29	28	26	49	39	(−7	−36	−45	−54)	36
1.0 (nominal)				5.00	1.7		54	39	35	34	32	53	49	(−3	−26	−30	−32)	43
Gas weld deposit; stabilized 24 hr. at the test temperature before testing.																		
1.0 (nominal)				5.00	1.7		57	40	34	32	30	58	56	8	−19	−26	−32	52

NOTES: (a) indicates room temperature hardness after test at indicated temperature; (b) indicates room temperature hardness before test at indicated temperature.
Numerals (0, 1, 2, 4) indicate loading interval in minutes.
Rockwell "C" values in parentheses indicate that they were converted from RA readings, using the chart of Fig. 3.
Air cooling also followed the stabilization treatments.

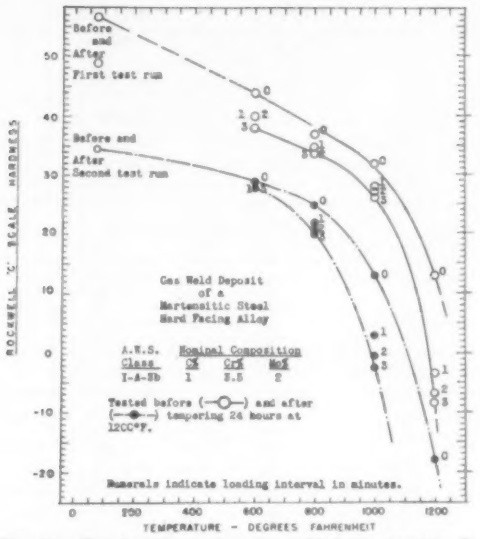

Before and After First test run
Before and After Second test run
Gas Weld Deposit of a Martensitic Steel Hard Facing Alloy
A.W.S. Class I-A-3b
Nominal Composition C% 1 Cr% 3.5 Mo% 2
Tested before (—○—) and after (—●—) tempering 24 hours at 1200°F.
Numerals indicate loading interval in minutes.
ROCKWELL "C" SCALE HARDNESS
TEMPERATURE - DEGREES FAHRENHEIT

Fig. 13 Hot hardness of a gas welded martensitic die steel hard-facing alloy before and after tempering 24 hr. at 1200° F.

are fabricated by filling mild steel tubing with mixtures of alloys such as ferrochromium, ferromanganese, etc. They sometimes give erratic results[3] as alloying is in the hands of the welder.

Five Per Cent Chromium Steels

Chromium steels like AISI 501 and 502 (4–6% Cr) and A.S.T.M. designation A157–44 types are widely used in refinery still tubes and parts, with carbon below 0.20%. They readily form martensitic structures when air-cooled from red heat. Hard-facing alloys on this base usually have higher carbon contents than the wrought and cast grades. Some indication of carbon effects is given in Table 4. Some molybdenum and nickel are frequently present in these compositions. The effect of temperature, before and after tempering at 1200° F., is shown in Fig. 12.

Medium Alloy Steels: A.W.S. Class I-A-3

The 3d Ed. (1950) Handbook classification designates the types with alloying elements between 6 and 12% (excluding carbon) as medium alloy steels, with subgroups (*a*) (0.20–0.60% carbon); (*b*) (0.60–1.70% carbon) and (*c*) (1.7–4.5% carbon). The last range covers alloy cast irons. Type I-A-3a thus covers the 0.21% C and 0.30% C steels in Table 4, and Type I-A-3b the higher carbon grades of Figs. 12 and 13. They are so soft at 1200° F. that they probably should be considered unsuitable for wear resistance with sustained loads at temperatures above 1000° F. At 1000° F., the effect of carbon is marked.

Several tests have been made on 9% chromium steel similar otherwise to the 0.21% C, 5% Cr steel in Table 4. The results were so nearly like those for 5% Cr that the data are not tabulated here.

Figure 13 describes a somewhat different steel that falls in this group only because of the usual percentages of manganese and silicon that it contains. Chromium and molybdenum nominally total only 5½%, placing it on the low alloy extreme of this classification group. Nevertheless, its properties are not much different from the higher alloy composition of Fig. 12 at 1000 and 1200° F. This steel is used for the facing and the repair of dies. In service it is frequently subjected momentarily to high temperatures and performs well under these conditions.

Medium High-Alloy Steels: A.W.S. Class I-A-4

This group covers hardenable iron-base alloys containing from 12 to 30% of elements other than iron and carbon, but excluding high-speed steels. Because of the hardenability stipulation it excludes austenitic steels, and thus tends to confine most of the commercial hard facings to the lower side of the range, since large amounts of most of the alloy additions, such as manganese, nickel, molybdenum, etc., will so stabilize austenite that weld deposits will not transform upon cooling.

The 12% Cr stainless steels are on the border of this group and are frequently used as the basis of hard-facing rods. Chromium in the 11–14% range renders these steels air-hardening, whereas sometimes higher percentages may prevent the formation of austenite (a 0.30% C–28% Cr alloy is ferritic at usual austenitizing temperatures) or if associated with nickel may so stabilize austenite that the hardening transformation does not occur. The 18% Cr – 8% Ni stainless steels are an example of the latter. The 12% chromium steels as a type tend to produce martensitic weld deposits.

This level of chromium (and higher percentages also) makes formation of special carbides of the Cr_7C_3 type possible. Secondary hardening in the neighborhood of 900° F. (see Fig. 7) is attributed to the precipitation of these. While they strengthen the alloy, they also confer brittleness at room temperature.

The 12% Cr steels are frequently employed for elevated temperature service up to perhaps 1200° F. Since the hardness of martensite is lost by tempering in this range it is customary to stabilize the structure by intentional tempering, usually at about 50° F. above the service temperature, before use. This explains the heat treatments that appear with the data in Table 5. Table 6 covers a similar steel tested as-cast, except for the stabilization treatments at each test temperature. These data are more representative of the expectancy for air-cooled weld deposits. Basically, this is a stainless steel of the cutlery type. While as-welded hardness may be high, better abrasion resistance is expected (with a sacrifice in toughness) at higher carbon levels.

The high-carbon, high-chromium tool steels are another variation on the 12% Cr base. These usually contain molybdenum to reinforce the air-hardening

Table 5—Hot Hardness of a Medium Carbon 12% Cr Steel

A.W.S. Class I-A-4b

Heat No.	C	Mn	Si	Cr
	%			
46-059	0.33	0.49	0.47	13.08

Specimens taken from small castings.
Individual specimens tested at each temperature.
Heat treated by tempering 6 hr. at 1350° F. before hot testing.
Room temperature hardness: C 24 before and C 22 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. (RC)	800° F. RA	1000° F. (RC)	1000° F. RA	1200° F. (RC)	1200° F. RA
0	15	9	54	− 7	46	−36	30
1	15	6	52	−20	38	−73	10
2	15	5	52	−24	36	−83	4
3	14	4	51	−27	34	−87	2

Individual specimens tested at each temperature.
Heat treated by hardening (1850° F.—2 hr.—air cool) and tempering 6 hr. at 1350° F.
Room temperature hardness: C21 before and C21 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. (RC)	800° F. RA	1000° F. (RC)	1000° F. RA	1200° F. (RC)	1200° F. RA
0	12	8	54	− 8	45	−38	28
1	12	6	52	−22	37	−75	8
2	11	6	52	−26	35	−88	2
3	11	5	52	−28	34	−93	−1

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

properties conferred by chromium. Such steels can be heat treated to a very high-hardness level (Rockwell C67), but they are relatively brittle when hard. Carbon content is so high that they also can be typed as cast irons, especially when they are used as castings or weld deposits. The hot working of wrought products confers some toughness by breaking up the continuity of free carbides, but brittle structures dependent on carbide segregation are the rule in weld deposits. However, the alloys are tough enough for die service, and as such are likely to be subjected to only momentary contacts with hot metal. The hot-hardness data were therefore taken on hardened specimens rather than after sustained tempering (Table 7).

Table 7—Hot Hardness of a High-Carbon 12% Cr Steel

A.W.S. Class I-A-4c

Heat No.	Chemical analysis, % C	Mn	Si	Cr	Mo
49-088	1.46	0.63	0.50	12.52	1.02

Specimens taken from small castings.
Individual specimens tested at each temperature.
Air hardened (1650° F.—2 hr.—air cool) before testing.
Room temperature hardness C61 before and C41 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	53	49	39	−14	42
1	51	47	30	−64	15
2	51	46	26	−78	7
3	50	46	25	−90	0

Individual specimens tested at each temperature.
Air hardened (2000° F.—2 hr.—air cool) before testing.
Room temperature hardness: C59 before and C46 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	42	38	29	− 2	48
1	39	36	25	−39	28
2	38	36	24	−50	22
3	38	36	23	−56	19

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

High-Speed Steels: A.W.S. Class I-A-5

Most hard-facing steels containing from perhaps 15 to 30% of special alloy elements fall in the high-speed tool class, and since these are logical steels to select for good hot hardness up to 1200° F. they have been segregated from the other high-alloy compositions.

Bishop and Cohen,[2] in their paper describing the apparatus from which the model used in this research was copied, provide data at 1000°, 1050° and 1100° F. These are reproduced in Table 8 and Fig. 14, for comparison with data to follow. The authors were interested in hot hardness because of a possible correlation with cutting abilities. This same inference is pertinent to

Table 6—Hot Hardness of Cast 12% Cr Steel

A.W.S. Class I-A-4b, Alloy Casting Institute Type HA32

Heat No.	C	Mn	Si	Cr
	%			
44-398	0.32	0.66	0.98	11.9

As cast specimens individually tested after stabilizing 24 hr. at the indicated test temperature. This steel also appears in Figs. 4 and 8.
Room temperature hardness: C52.5–54.2; average C53.2.

Loading interval, min.	Rockwell hardness R. T. RC	200° F. RC	400° F. RC	600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	53.2(a)	50.8(a)	49.9(a)	47.2(a)	45.4(a)	14.5(a)	31.5(a)	32.0(a)
1	52.7(a)	50.0(a)	49.1(a)	45.7(a)	44.4(a)	−1.5(a)	...	7.3(a)
2	52.4(a)	49.7(a)	48.8(a)	45.4(a)	44.1(a)	−5.5(a)	...	−0.4(a)
3	52.4(a)	49.7(a)	48.6(a)	45.1(a)	43.9(a)	−8.2(a)	...	−4.7(a)
4	53.6	51.0	49.7	45.3	44.0	−12.2	...	−6.5
6	53.3	51.0	49.7	45.3	42.8	−14.4	...	−9.0
8	53.3	51.0	48.8	45.3	41.5	−17.2	...	−13.0
12	53.3	51.0	48.6	44.6	40.3	−19.7	...	−17.0
16	53.3	51.0	48.6	44.4	40.1	−20.0	...	−21.0
20	53.2	51.0	48.6	45.4	40.6	−20.0	...	−23.0
24	53.2	51.0	48.6	44.8	41.0	−20.0	...	−25.0

NOTE: Several determinations were averaged for the 0-, 1-, 2- and 3-min. values, as indicated by (a), but the longer time data are based on a single test run. This accounts for an apparent discrepancy between some of the 3- and 4-minute values.
(RC) in parentheses indicates that the Rockwell "C" value in that column is converted from Rockwell "A" determinations.

Table 8—Hot Hardness of Various Wrought High-Speed Steels as Reported by Bishop and Cohen,

Steel type	Composition C	W	Mo	Cr	V	Co	Heat treatment, °F. Harden	Temper	Hardness at 70° Hardened	Tempered	At end*	—Hot hardness, °F.— 1000	1050	1100
14-4-1	0.71	14.60	..	4.13	2.08	...	2300	1050	66.0	64.0	63.3	58.0	56.5	54.5
18-4-1	0.72	18.16	..	4.05	1.04	...	2350	1050	66.0	65.2	64.8	58.3	57.2	55.8
18-4-2	0.835	18.52	0.65	4.24	2.12	...	2350	1050	66.1	66.3	66.2	60.2	57.8	56.2
W-5% Co	0.725	17.29	0.52	4.00	1.06	4.73	2400	1050	65.0	66.2	65.8	60.2	59.5	58.7
W-12% Co	0.795	20.38	0.65	4.22	1.64	12.44	2375	1000	64.0	67.2	67.1	62.0	61.4	59.6
6-6-4-2	0.855	6.50	5.62	4.48	2.08	...	2235	1050	65.1	65.3	64.8	58.0	57.2	55.9
Mo-V	0.845	...	8.29	4.12	2.24	...	2215	1025	64.9	65.3	64.8	58.0	57.0	55.5
High C., high V	1.275	5.71	4.47	4.44	4.15	...	2220	1075	66.0	65.4	65.1	59.0	57.5	56.7

* After hot-hardness tests. All hardness values an Rockwell "C" scale.

hard facings as some of them are used for the edges of shear blades and other tools for hot cutting.

Wrought high-speed steel tools receive a complex hardening and tempering treatment well described by Cohen and Gordon.[19, 20] Table 9 contains data on such a material, heat treated as in current industrial practice. In contrast, Table 10 and Fig. 15 refer to weld deposits, which are expected to contain some retained austenite. Note that the weld deposit hardness at 1000° F. is below that of the heat treated wrought specimens. This confirms the conclusion that maximum transformation of residual austenite is desirable for hot hardness.[2] Such could be attained by heat treatment of the weld deposits, but since one outstanding merit of the hard-facing technique is simplicity the complications of further treatment are generally avoided in practice.

The excellent hot hardness and creep resistance of the high-speed steels at 1000° F. is largely lost at 1200° F.

Table 9—Hot Hardness of Wrought 18% W–4% Cr–1% V High-Speed Steel

A.W.S. Class I-A-5

Nominal Composition, % C	Mn	Si	Cr	W	V
0.70	0.25	0.25	4.0	18.0	1.0

Heat treatment: 1600° F.—2 hr.—furnace cool (annealing). 1050° F.—1 hr.—transferred to 1650° F.—40 min.—transferred to 2350° F.—15 min.—oil quenched. Then tempered twice by 1050° F.—2 hr.—air cool.

Consecutive tests at four temperatures on one specimen.

Room temperature hardness: C64 (Bhn. 682–698) before hot testing and C58 after the first hot test run.

Loading interval, min.	Rockwell "C" scale hardness; 600° F.	800° F.	1000° F.	1200° F.
0	59	57	54	41
1	58	55	53	34
2	58	54	52	33
3	57	54	52	31

Consecutive tests on the same specimen used above, but tempered 24 hr. at 1200° F., following the first test run.

Room temperature hardness: C49 before and C50 after second hot test run.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	RA
0	44	40	34	16	58
1	43	37	25	2	50
2	43	36	23	−1	49
3	43	36	22	−4	47

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

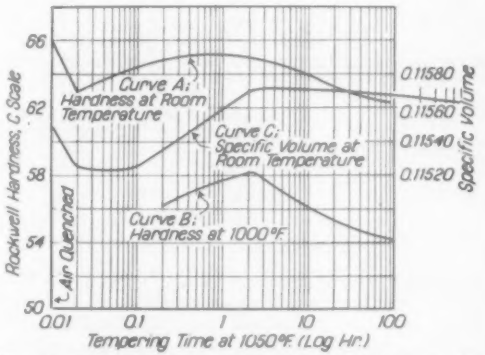


Fig. 14 The effect of tempering time at 1050° F. on the specific volume and on the hardness at room temperature and 1000° F. of 18% W—4% Cr—1% V high-speed steel as reported by Bishop and Cohen[2]

The rising specific volume curve reflects austenite transformation and indicates that maximum hot hardness occurs at the point where transformation is complete

Like other iron-base alloys that depend on martensitic hardening, this is probably the maximum temperature for general usefulness. For service above 1200° F., some other material should be selected.

Alloy Cast Irons; A.W.S. Class I-A-2d with 1–6% Alloys; A.W.S. Class I-A-3c with 6–12% Alloys; A.W.S. Class I-A-4d with 12–30% Alloys

Since carbon has a potent influence on properties the irons with carbon between 1.5% and perhaps 4.5% are discussed together here. Engineering gray cast irons, which are graphitized largely through the influence of silicon, have little use as hard facings, though they are occasionally appropriate as bearing surfaces. Elements like chromium and manganese tend to prevent graphitization and the castings or air-cooled weld deposits from low-alloy irons containing them are characteristically hard, with a white fracture. Usually their structure is a mixture of pearlite and cementite. Though used to some extent for hard facing, the small extra cost of the proper alloy combination to produce martensitic irons suggests that the pearlitic irons are uneconomical.

Martensitic irons can be made with a wide range of compositions. A low-alloy representative is Ni-Hard,[21] with a nominal alloy content of 4.5% Ni and 1.5% Cr. With about 3.5% carbon, this combination produces a

Table 10—Hot Hardness of High-Speed Steel Weld Deposits

A.W.S. Class I-A-5

Nominal composition, %				
C	*Cr*	*Mo*	*W*	*V*
0.8	4.0	9.0	1.5	1.0

Determinations made on individual specimens, each stabilized 24 hr. at the test temperature before testing.

Room temperature hardness: C60 before and C62 after testing at 1000° F., C57 before and C56 after testing at 1200° F.

Loading interval, min.	*Rockwell hardness; gas weld deposit; 1000° F. RC*	*1200° F. RC*
0	45	23
1	41	2
2	40	−3
4	39	−9

Room temperature hardness: C55 before and C54 after testing at 1000° F., C40 before and C31 after testing at 1200° F.

Loading interval, min.	*Rockwell hardness; arc weld deposit. 1000° F. RC*	*1200° F. RC*
0	46	3
1	45	−27
2	44	−30
4	43	−32

material consisting chiefly of hard cementite and austenite that has mostly transformed to martensite. It has developed an enviable record for abrasion resistance in applications where its brittleness is not a handicap. Hot-hardness data appear in Fig. 16 for castings and in Fig. 17 for weld deposits. They suggest that this alloy should not be used above 800° F. because of the softening that occurs.

While the strong carbide-forming elements are not so efficient as nickel in producing a martensitic structure (in cast irons) upon air cooling, they contribute greater

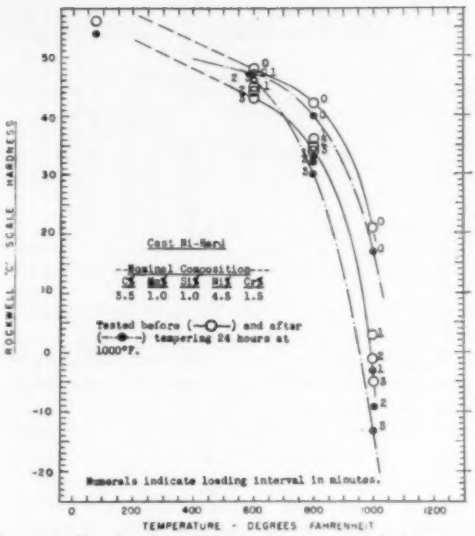

ROCKWELL "C" SCALE HARDNESS
Cast Bi-Metal
--Nominal Composition--
C% Mn% Si% Ni% Cr%
3.5 1.0 1.0 4.5 1.5
Tested before (—○—) and after (—●—) tempering 24 hours at 1000°F.
Numerals indicate loading interval in minutes.
TEMPERATURE - DEGREES FAHRENHEIT

Fig. 16 Hot hardness of martensitic nickel-chromium iron from small castings, showing the effect of tempering at 1000° F.

hot hardness. Figures 18 and 19 describe chromium-molybdenum and chromium-tungsten martensitic irons, respectively. The chromium-tungsten iron is related to the high-speed steels in alloy content and like them has outstanding hardness around 1000° F.

These two alloys are not greatly affected by tempering, the chief result being loss of creep resistance at 1200° F. While both can be used up to this temperature it is suggested that service up to 900° F. for the chromium-molybdenum type or up to 1100° F. for

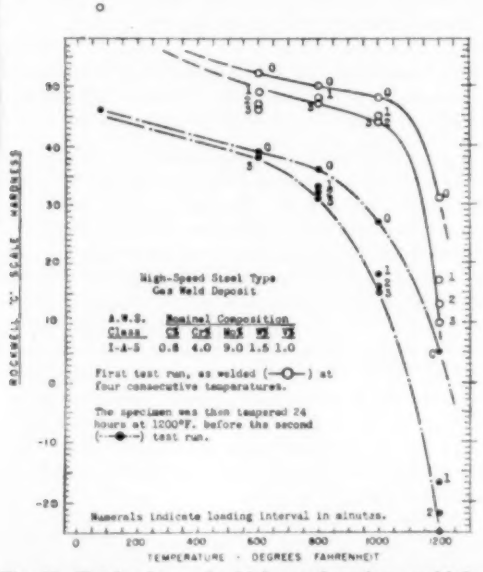

ROCKWELL "C" SCALE HARDNESS
High-Speed Steel Type
Gas Weld Deposit
A.W.S. Class I-A-5
Nominal Composition
C% Cr% Mo% W% V%
0.8 4.0 9.0 1.5 1.0
First test run, as welded (—○—) at four consecutive temperatures.
The specimen was then tempered 24 hours at 1200°F. before the second (—●—) test run.
Numerals indicate loading interval in minutes.
TEMPERATURE - DEGREES FAHRENHEIT

Fig. 15 Hot hardness of a high-speed steel gas weld deposit before and after tempering 24 hr. at 1200° F.

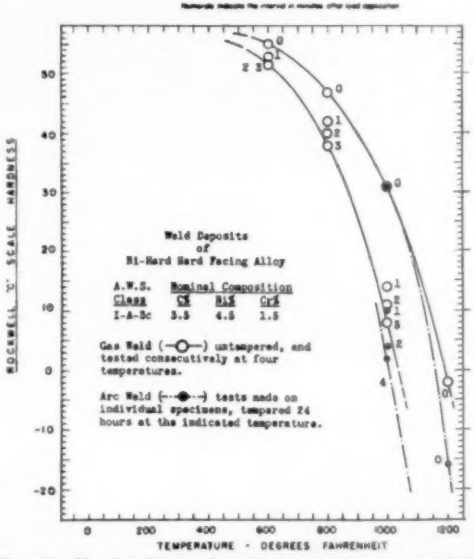

ROCKWELL "C" SCALE HARDNESS
Weld Deposits
of
Bi-Hard Hard Facing Alloy
A.W.S. Class I-A-Bc
Nominal Composition
C% Ni% Cr%
3.5 4.5 1.5
Gas Weld (—○—) untempered, and tested consecutively at four temperatures.
Arc Weld (—●—) tests made on individual specimens, tempered 24 hours at the indicated temperature.
TEMPERATURE - DEGREES FAHRENHEIT

Fig. 17 Hot hardness of weld deposits from a nickel-chromium martensitic iron

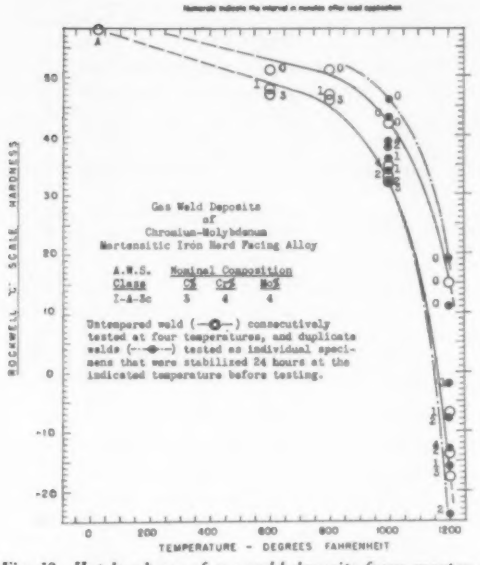


Fig. 18 Hot hardness of gas weld deposits from martensitic Cr-Mo iron: tested with and without tempering before determining hardness

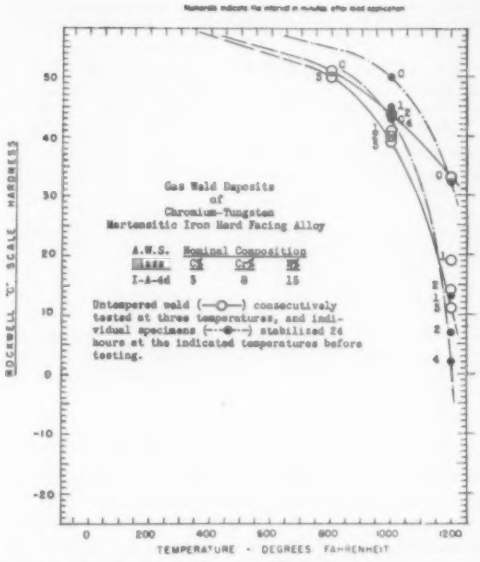


Fig. 19 Hot hardness of gas weld deposits from martensitic Cr-W iron: tested with and without tempering before determining hardness

the chromium-tungsten type is appropriate. At and above 1200° F., the cobalt-base alloys are to be preferred.

The chromium-tungsten alloy of Fig. 19 is a very useful hard-facing material as it combines high hardness (up to C65 for gas welds), excellent abrasion resistance, high compressive strength and excellent hot hardness. Under high-stress abrasion, as produced by the American Brake Shoe Laboratory apparatus, gas welds consistently give wet quartz sand abrasion factors in the vicinity of 0.30–0.35. This average performance is better than that of any other alloy covered by this survey with the exception of the composite tungsten carbide types. Under low stress abrasion, as produced experimentally by Haworth's rubber wheel apparatus,[22] it also is outstanding.

The hot-hardness range is comparable to that of high-speed steel except that it seems to be less sensitive to tempering at 1200° F. When estimating potential hot-wear resistance it should be remembered that the iron contains a relatively large proportion of massive hard carbides in comparison with the steels. Since the hardness values suggest that the matrices probably are similar in strength the greater carbide volume should provide better wearing qualities. Though this pattern is hypothetical for hot wear, because a good laboratory test is lacking, it has been identified in room temperature testing with the American Brake Shoe Laboratory apparatus and by Haworth[22] in testing a series of alloys containing carbon and chromium as variables.

The choice between high-speed steel and the chromium-tungsten iron should be made on the basis of the toughness required in service. High-speed steel is expected to be tougher but since it is a relatively brittle steel the difference is not great. If the iron will serve, better all-around performance may be expected from it.

Austenitic Stainless Steels: A.W.S. Class I-B-1a—Low (Less Than 0.20%) Carbon; A.W.S. Class I-B-1b High (0.20–1.7%) Carbon, Low (Less Than 7%) Nickel; A.W.S. Class I-B-1c—High (0.20–1.7%) Carbon, Chromium Above 15%, Nickel Above 7%

The low-carbon austenitic chromium and chromium-nickel steels, as represented by the variety of wrought compositions available as sheet, tubing, bar stock and welding rod, are generally employed either as welded overlays for corrosion resistance, including hot gas corrosion, or as filler metals for fabrication welds. For such purposes hot hardness is a minor consideration. Even at room temperature hardness is low unless it has been raised by cold work.

For elevated temperature service hot tensile and creep properties may be important, especially if the welds are made on equally competent materials. However, these data are too readily available in engineering literature to justify repetition here. It should suffice to note that low hardness is no index of creep resistance, which for chromium-nickel-iron alloys is good in comparison with many harder steels, especially above 1200° F. Hot gas corrosion resistance depends largely on the chromium and nickel contents (Fig. 10) and can be estimated reasonably well from the results of Alloy Casting Institute research.[16–18]

The range from 0.20–0.70% carbon contains most of the important cast heat-resistant alloys. Because of the similarity between cast structures and weld deposits,

Table 11—Hot Hardness of 26% Cr – 12% Ni Heat-Resistant Alloy

A.W.S. Class I-B-1c, Alloy Casting Institute Type HH32

Heat No.	% C	Mn	Si	Ni	Cr	N
CH518	0.32	0.46	0.45	11.5	25.9	0.16

Individual specimens, tested as cast.

Room temperature hardness: C8 before and C12 after testing at 1200° F.

Loading interval, min.	Rockwell hardness. 600° F. (RC)	RA	800° F. (RC)	RA	1000° F. (RC)	RA	1200° F. (RC)	RA
0	−9	44.4	−11	43.3	−15	41.0	−21	37.9
1	−9	44.2	−13	41.8	−17	40.0	−23	36.7
2	−9	44.0	−15	41.0	−18	39.5	−24	36.1
3	−9	43.9	−16	40.5	−18	39.2	−25	35.6

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

Table 13—Hot Hardness of 26% Cr – 35% Ni Heat-Resistant Alloy

A.W.S. Class I-B-1c, Alloy Casting Institute Type HP40

Heat No.	% C	Mn	Si	Ni	Cr	N
45-705	0.40	0.95	1.34	34.9	25.8	0.07

Individual specimens, tested as-cast.

Room temperature hardness: (C3) before and (C3) after testing at 1200° F.

Loading interval, min.	Rockwell hardness. 600° F. (RC)	RA	800° F. (RC)	RA	1000° F. (RC)	RA	1200° F. (RC)	RA
0	−11	43.3	−13	41.9	−17	39.7	−25	35.6
1	−12	42.7	−14	41.7	−18	39.5	−28	33.8
2	−12	42.6	−14	41.6	−18	39.3	−29	33.4
3	−12	42.6	−14	41.5	−19	38.9	−29	33.2

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

neither of which is usually hot worked after solidification, data from castings are pertinent to welded overlays of similar composition. If nickel is below 10% the structure is ordinarily not stably austenitic; thus a representative of the I-B-1b class is unlikely to be encountered.

A ferritic grade, the A.C.I. "HC" type, is the low-nickel end of a series containing about 26% chromium. Both ferrite and austenite, as well as carbides and sometimes the sigma phase,[24] appear in the structures of this series, the proportions changing until at about 0.30% carbon and 12% nickel or 0.20% carbon and 20% nickel the alloys may become wholly austenitic.[25–27] Ferrite is weak and ductile at high temperatures and thus increasing nickel is associated with increasing hot strength. The real justification for using unstable or borderline members of this transition series is based on scaling resistance in reducing atmospheres containing considerable sulphur. If this element, either as hydrogen sulphide or in elemental form, exceeds perhaps two or three grains per cubic foot the scaling of nickel-bearing alloys is accelerated. Otherwise, 26% Cr gives good surface protection up to 2000° F. Table 11 describes a wholly austenitic 26% Cr – 12% Ni alloy of the A.C.I. "HH" type to represent the upper end of the transition series. Except for extreme resistance to hot sulphur attack this is about the lowest alloy content that would be logically selected for service above 1600° F.

For hot wear resistance, higher carbon contents seem indicated, as they produce hard massive carbide particles in the austenitic matrix. However, there is no good experimental evidence to prove this correlation with hot wear. The conclusion is based largely on inference. Data for a high carbon "HH" alloy appears in Table 12. Note that while the hardness level of these materials is low, the creep resistance, as shown by the small difference between the 0- and 3-minute values, is relatively high at 1200° F. This advantage over the ferritic and martensitic alloys increases above 1200° F. This statement is based on creep rather than hot hardness testing, but considerable evidence not pertinent to this paper indicates that for most applications above 1200° F., where hot strength is required, an austenitic structure should be selected. The creep properties of the alloy in Table 11 have been described in detail in another publication.[26]

Tables 13 and 14 pertain to other stably austenitic

Table 12—Hot Hardness of High-Carbon 26% Cr – 14% Ni Alloy

A.W.S. Class I-B-1c, Alloy Casting Institute Class HH52

Heat No.	% C	Mn	Si	Ni	Cr	N
48-840	0.52	0.86	0.92	13.8	26.4	0.12

As-cast specimen tested successively at four temperatures.

Room temperature hardness: C23 before and C23 after hot tests.

Loading interval, min.	Rockwell hardness. 600° F. (RC)	RA	800° F. (RC)	RA	1000° F. (RC)	RA	1200° F. (RC)	RA
0	10	54.2	7	52.7	3	50.7	−3	47.4
1	9	53.9	6	52.0	2	50.3	−7	45.4
2	9	53.8	5	51.9	2	50.2	−8	44.8
3	9	53.7	5	51.9	2	50.2	−9	44.3

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

Table 14—Hot Hardness of 18% Cr – 38% Ni Heat-Resistant Alloy

A.W.S. Class I-B-1c, Alloy Casting Institute Type HU47

Heat No.	% C	Mn	Si	Ni	Cr	N
48-249	0.47	0.67	1.23	38.2	17.5	0.08

Individual specimens, tested as-cast.

Room temperature hardness: (C8) before and (C9) after testing at 1200° F.

Loading interval, min.	Rockwell hardness. 600° F. (RC)	RA	800° F. (RC)	RA	1000° F. (RC)	RA	1200° F. (RC)	RA
0	−4	47.0	−5	46.5	− 8	44.9	− 8	44.6
1	−5	46.5	−6	45.8	− 9	44.0	−14	41.2
2	−5	46.4	−7	45.5	−10	43.4	−17	40.1
3	−5	46.3	−7	45.1	−11	42.9	−17	39.8

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

chromium-nickel-iron alloys. All of the higher carbon heat-resistant alloys are not stocked commercially as welding rod since the bare wire is usually made low in carbon for easy hot working. Higher carbon arc welds are readily made by including carbon in the flux coating and such electrodes can be obtained from welding rod producers. For gas welding the rods are generally cast and used by foundries in connection with high-alloy castings.

A research program dealing with the welding of these high alloy heat-resistant metals is currently being undertaken by the Welding Research Council in cooperation with the Alloy Casting Institute.

Austenitic Manganese Steels; A.W.S. Class I-B-2

Austenitic manganese steels[28] are usually selected where service requires both wear resistance and toughness. The tough structure transforms in the 600–1200° F. range of this hot-hardness survey, thereby destroying the toughness for which the material is selected. Even the heat of welding has an adverse effect.[29] This loss of toughness is shown in Fig. 20,[30] which justifies the conclusion that 12–14% manganese steel (the Hadfield type) should not be employed at the temperatures being considered here. The modification with carbon below 1% and with several per cent of nickel is slightly less susceptible to this effect, but the difference is not significant in practice.

Several tests on arc weld deposits of rods intended for welding manganese steel gave converted Rockwell "C" ranges of −15 to −35 (0- to 4-min. loading) at 1000° F. and −35 to −64 (0- to 2-min. loading) at 1200° F. These data and the embrittlement that occurs should discourage the use of this material at these temperatures.

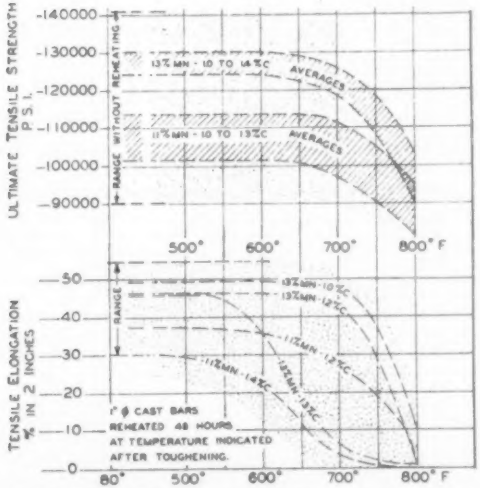


Fig. 20 Embrittlement from reheating austenitic manganese steel

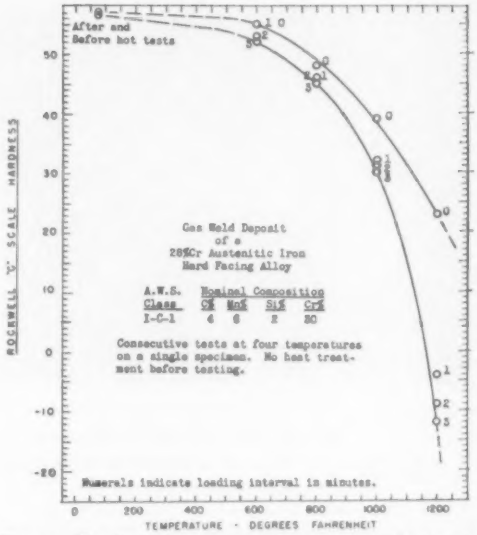


Fig. 21 Hot hardness of an austenitic gas weld deposit from a 28% chromium iron

The effect of tempering during the test run was negligible

Austenitic Cast Irons: A.W.S. Class I-C-1—22–40% Cr, Carbon Above 2%; A.W.S. Class I-C-2a—4–10% Cr with 7–12% Total Alloys, Carbon Above 1.7%; A.W.S. Class I-C-2b—Cr Under 14% with 12–30% Total Alloys, Carbon Above 2.5%; A.W.S. Class I-C-2c—30–50% Total Alloys, Carbon Above 2.0%

The I-C-1 class covers a widely used welding rod type containing approximately 4% carbon, 6% manganese,

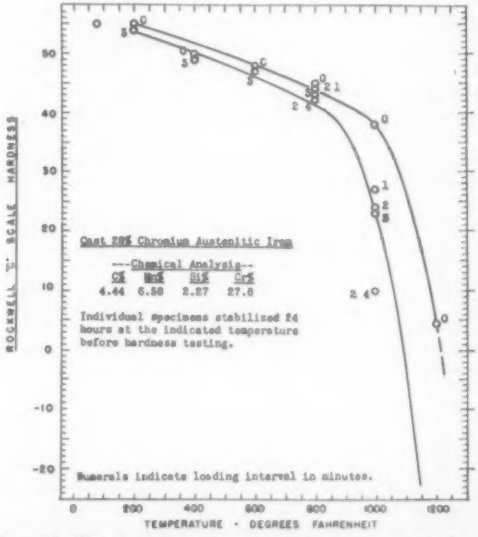
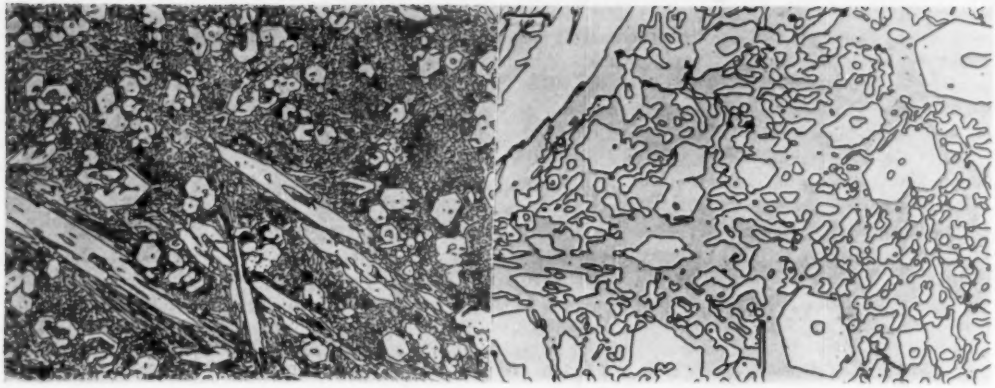


Fig. 22 Hot hardness of a cast 28% Cr austenitic iron



(A) (B)

Fig. 23 Structure of gas weld deposit from 4% C—6% Mn—2% Si—28% Cr welding rod

(A) Hardness, Rc, 58; Bhn. 532; subsequent hot-hardness data appear in figure. 100 ×

(B) Wet quartz No. 55 sand abrasion factor 0.78 vs. 1.00 for S.A.E. 1020 steel. 500 ×. (See "Hard Surfacing by Fusion Welding" for test description and comparative data.)

2% silicon and 30% chromium; and several 22-30% chromium cast irons used more for heat- and abrasion-resistant castings than for welding. The HC-250 iron (Alloy Casting Institute designation system) of Fig. 6 is a representative of the latter. The manganese-silicon-chromium welding rod is popular for hard-facing agricultural tools. The effect of temperature on an untempered gas weld deposit appears in Fig. 21. Since the structure is initially austenitic the effect of sustained tempering need not be a matter of concern. In the series of tests of this alloy type plotted in Figs. 4, 8 and 22 the entire spread of room temperature hardness values after heat treating and testing from 200 to 1200° F. was Rockwell C55 ± 2.5. This range is attributed largely to the heterogeneous structure, which is illustrated in Figs. 23 (*A*) and (*B*). In the austenite matrix are pseudohexagonal carbide spines and crystals of the Cr_7C_3 type. These differ considerably in hardness, the austenite being about 440 VPN and the large carbide crystals about 2100 VPN when tested with a Vickers diamond indentor under a 25-gm. load in a Tukon instrument at room temperature. Figures 21 and 23 refer to the same weld deposit.

Austenitic irons of the I-C-2a and I-C-2b classes have not yet been included in this program. The higher alloy I-C-2c class is represented by Tables 15, 16 and 17. While these are relatively soft they have good creep resistance and thus at 1000 and 1200° F.

Table 15—Austenitic 29% Cr – 9% Ni Heat-Resistant Cast Iron

A.W.S. Class I-C-2c, Alloy Casting Institute Type HE250

Heat No.	Chemical analysis, % C	Mn	Si	Ni	Cr	N
47-518	2.50	0.72	1.53	9.0	28.5	0.08

Individual specimens from a small casting tested at each temperature without previous heat treatment.

Room temperature hardness: C35 before and C35 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	24	22	17	8	53
1	24	21	16	2	50
2	23	21	15	0	49
3	23	20	15	−2	48

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

Table 16—Austenitic 26% Cr – 12% Ni Heat-Resistant Cast Iron

A.W.S. Class I-C-2c, Alloy Casting Institute Type HH250

Heat No.	% C	Mn	Si	Ni	Cr	N
47-528	2.55	0.76	1.55	12.0	25.7	0.09

One specimen from a small casting tested consecutively at four temperatures without previous heat treatment.

Room temperature hardness: C36 before and C36 after testing.

Loading interval min.	Rockwell "C" scale hardness 600° F.	800° F.	1000° F.	1200° F.
0	26	24	19	14
1	26	24	16	6
2	25	24	16	4
3	25	23	15	3

Table 17—Austenitic 26% Cr – 20% Ni Heat-Resistant Cast Iron

A.W.S. Class I-C-2c, Alloy Casting Institute Type HK250

Heat No.	% C	Mn	Si	Ni	Cr	N
47-577	2.42	0.80	1.63	19.9	25.6	0.09

Individual specimens from a small casting tested at each temperature without previous heat treatment.

Room temperature hardness: C31 before and C30 after testing.

Loading interval, min.	Rockwell hardness 600° F. RC	800° F. RC	1000° F. RC	1200° F. (RC)	1200° F. RA
0	23	24	19	9	54
1	23	23	18	2	50
2	22	23	17	1	49
3	22	23	17	−2	48

NOTE: (RC) in parentheses indicates that the Rockwell "C" values in that column are converted from Rockwell "A" determinations.

they surpass many of the martensitic steels that are much harder at room temperature.

Cobalt Base Alloys: A.W.S. Class II-A—Low Alloy; A.W.S. Class II-B—High Alloy

The outstanding alloys for hot hardness above 1200° F. contain 25% or more of chromium, several per cent of tungsten and from 0.5 to 3.0% carbon. The balance is chiefly cobalt, which ranges upward from about 43% to perhaps 65%. Unfortunately, this adds considerably to their cost.

The most popular variations are:

A.W.S. class	*Usual commercial designation*	*Nominal composition, %*			
		C	*Cr*	*W*	*Co*
IIA	"6"	1.0	28	4.5	55 min.
IIB	"1"	2.5	31	12.5	43 min.

These are merely two points in an extensive system of alloys that covers a wide range in properties, but they have been rather well standardized by a number of manufacturers and they represent focal points of considerable field experience.

At one end of the series that has received industrial attention is the X-40 alloy of the Office of Scientific Research and Development program directed toward improved gas turbine alloys. This material in the form of castings (XR No. 71—0.5% C, 0.6% Mn, 0.7% Si, 25% Cr, 10% Ni, 55% Co, 7% W, 0.6% Fe)* was about the strongest material at 1800° F. of the many machinable alloys included in this wartime research effort in which many laboratories participated.

At room temperature the hardness of the 0.5% C alloy is near 300 Brinell. At 1.0% carbon it approximates 400 Bhn. and at 2.5% C it is seldom under 500 Bhn. Hot hardness follows much the same pattern, with zero time values of Rockwell C27–C31 for 1% C and C35–C47 for 2.5% C at 1000° F.; C21–C27 for 1% C and C35–C44 for 2.5% C at 1200° F. Carbon is thus a potent minor component.

Note that these hardness values at 1200° F. are relatively high. This advantage is more marked when the effect of creep is considered; the cobalt-base alloys lose only a few points of hardness in 3 or 4 min. at 1200° F., while the other types that exhibit C30 or above at this temperature lose many scale points in the first three minutes. Figures 24 and 25 are temperature-hardness plots of gas and arc welds. Up to 1000° F. the effect of creep during a 3-min. interval is within the scatter band of experimental error, as is the difference in room temperature hardness before and after testing up through 1200° F. Thus creep and tempering effects can be considered negligible up to these values.

At 1200° F. three patterns were noticed in testing about a dozen gas and arc welds. Some deposits, like the 1.0% C arc weld in Fig. 25, held the 1000° F. type of creep resistance up to 1200° F. Others gave a relatively sharp drop between 1000° and 1200° F. like the 2.5% C arc weld of Fig. 25. The third pattern is intermediate between these two, like the two gas welds of Fig. 24. These variations are attributed to differ-

* Creep, stress-rupture and tensile data from 1200 to 1800° F. for this alloy are conveniently summarized in reference 31.

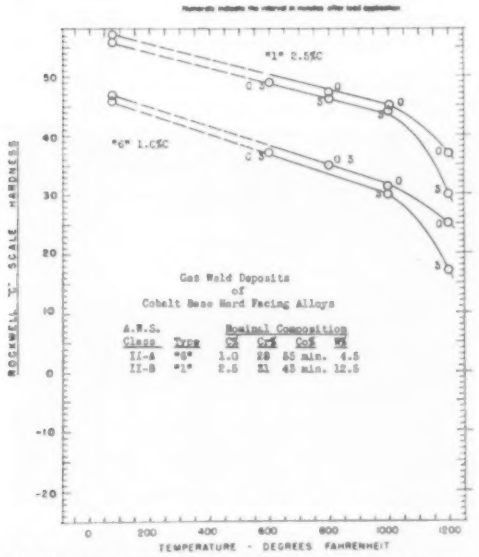


Fig. 24 Hot hardness of gas-welded overlays from industrially standardized cobalt base alloys

Successive tests at four temperatures on one untempered specimen of each type

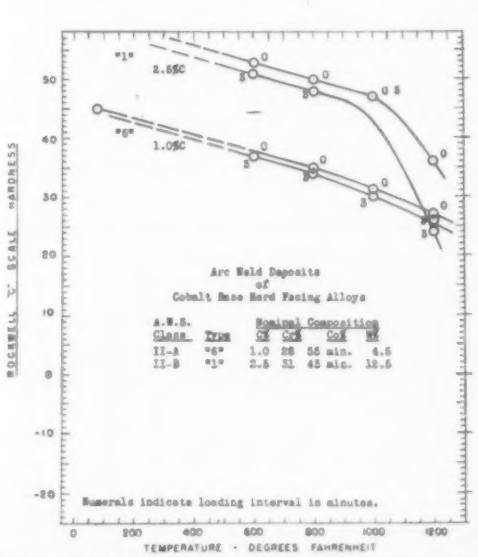


Fig. 25 Hot hardness of arc-welded overlays from industrially standardized cobalt base alloys

Successive tests at four temperatures on one untempered specimen of each type

Table 18—Hot-Hardness Survey of Cobalt Base Alloys

A.W.S. Class II-A, Usual Commercial Designation: No. 6

Nominal composition: 1.0% C, 28% Cr, 4.5% W, balance cobalt.

	Rockwell "C" scale hardness.									
	75° F.		*600° F.*		*800° F.*		*1000° F.*		*1200° F.*	
Specimen	*B*	*A*	*0*	*3*	*0*	*3*	*0*	*3*	*0*	*3*
Gas weld	47	45	37	37	35	35	31	30	25	17
Arc weld	45	45	37	37	35	34	31	30	27	26
Gas weld	44	44	..	..	34	33	31	30	27	25
Gas weld	41	42	..	..	31	30	27	26	21	14
Casting (st)	40	40	..	..	..	..	28	26	22	16
Av.	43	43	37	37	34	33	30	28	24	20

A.W.S. Class II-A, Usual Commercial Designation: No. 1

Nominal composition: 2.5% C, 31% Cr, 12.5% W, balance cobalt.

	Rockwell "C" scale hardness.									
	75° F.		*600° F.*		*800° F.*		*1000° F.*		*1200° F.*	
Specimen	*B*	*A*	*0*	*3*	*0*	*3*	*0*	*3*	*0*	*3*
Arc weld	59	59	53	51	50	48	47	47	36	24
Gas weld	57	55	50	49	47	46	45	44	37	30
Gas weld	53	53	..	..	47	46	44	43	44	42
Gas weld	55	56	..	..	48	47	46	45	42	40
Gas weld (st)	54	54	..	..	..	..	45	43	40	32
Arc weld (st)	51	51	..	..	..	..	35	32	35	23
Casting (st)	55	54	..	..	..	..	45	43	41	35
Av.	55	55	52	50	48	47	44	43	39	32

Code: B = before hot testing; A = after testing at 1200° F.; 0 and 3 indicate the interval in minutes after load application; (st) indicates that individual specimens were stabilized by tempering 24 hr. at the test temperature before testing.

ences in composition and structure as a result of segregation or dilution. They can occur for both carbon levels and both welding methods, but because of the likelihood of greater penetration and greater overlay dilution from base metal fusion they are more to be expected with arc welds. The undesirable pattern, with a marked drop in creep resistance at 1200° F., should be consciously minimized during welding by avoiding such dilution.

The ranges and averages of tests on several specimens of each of two compositions appear in Table 18. These are the most important of the hard-facing alloys with high hot hardness and they are the most generally available.

Much less important, but sometimes of interest for special applications, are modifications of the basic types. These usually are aimed at even higher hot hardness by an increase in tungsten content or addition of other strong carbide-forming elements such as molybdenum, columbium, tantalum, vanadium and the like. Table 19 contains several examples of these. The effectiveness of tungsten is demonstrated but the increase in hot hardness seems small in relation to the 20% increment in tungsten and seems difficult to justify economically.

Hot hardness alone does not sufficiently differentiate the 1.0 and 2.5% carbon types, since other qualities, notably toughness, control their selection. The 1% carbon alloy can be lathe turned with difficulty and subjected to some other machining operations, but the 2.5% carbon grade must be shaped by grinding.* It

Table 19—Hot Hardness of Modified Cobalt Base Alloys

A.W.S. Class II-B

Untempered gas weld deposits tested successively at three temperatures.

Nominal composition, %				*Rockwell "C" scale hardness*				
				75° F.	*Min.*	*800 ° F.*	*1000 ° F.*	*1200 ° F.*
C:	1.5	W:	13	Before hot test-	0	50	49	43
Cr:	28	Cb:	4	ing, 57; after	1	50	48	28
Co:	48	Cb:	2	56	2	50	48	20
					3	50	48	11
C:	2.0	.....		Before hot test-	0	49	48	41
Cr:	40	Mo:	10	ing, 57; after,	1	48	47	38
Co:	24	W:	1	57	2	48	47	38
					3	48	47	37
C:	2.8	.....		Before hot test-	0	37	35	30
Cr:	20	Mo:	8	ing, 43; after,	1	37	34	27
Co:	30	Ni:	16	43	2	37	34	26
					3	37	34	26
C:	2.3	W:	32	Before hot test-	0	56	53	48
Cr:	24	.....		ing, 60; after,	1	55	52	47
Co:	38	.....		60	2	55	52	46
					3	55	52	46

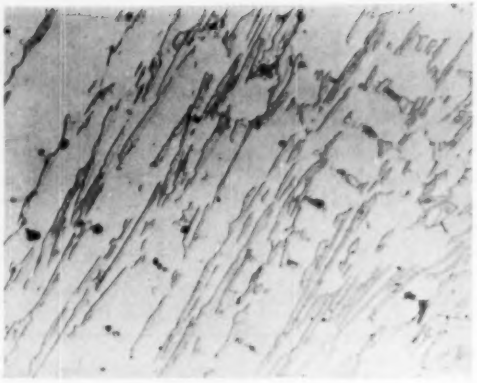


Fig. 26 Photomicrograph at 500 × of an arc weld deposit from a 1% C—28% Cr—60% Co—3.5% W welding rod

The structure consists of a moderate volume of carbides in a cobalt rich solid solution

* While the 2.5% C alloy has been machined at 0.006–0.010 in. feeds with cemented carbide tools, it is unlikely that this would be more economical than grinding.

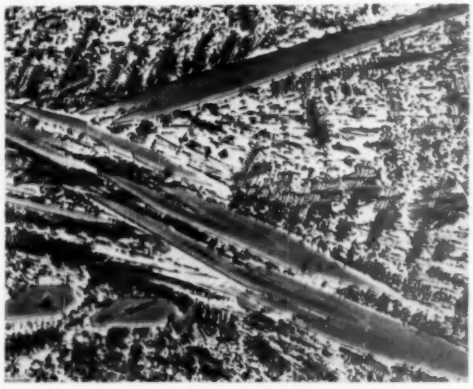


Fig. 27 Photomicrograph at 500 × of an arc weld deposit from a 2.6% C—32% Cr—43% Co—14% W welding rod

The structure consists of prominent pseudohexagonal spines of the Cr_7C_3 type in a matrix composed largely of an intimate mixture of finer carbides and solid solution. The "fishbone" structure indicates areas of eutectic composition. (Metal Congress prize winning photograph by R. J. Gray)

Table 20—Compression Test Properties of Cobalt Base Alloys

Chemical composition, %						Compression test data: Yield strength, psi. 0.05% set	0.10% set	Ultimate strength, psi.	Deformation, % Elastic	Plastic	Total	Hardness Rc	Bhn.
C	Mn	Si	Cr	Co	W								
Specimens 0.400 in. diameter x 1.00 in. high machined from small castings													
0.75	0.88	1.92	27.2	58.9	5.1	53,000	64,000	155,500	1.03	7.06	8.09	41	351
0.86	1.03	1.60	25.2	58.8	5.1	67,000	76,000	209,000	0.95	9.76	10.71	41	375
1.07	3.67	1.56	27.9	52.5	5.9	64,000	73,000	227,000	>.40	7.34+	7.74+	41	364
Specimens 0.400 in. diameter x 1.25 in. high machined from small castings													
0.75	0.88	1.92	27.2	58.9	5.1	70,000	76,000	174,000	0.58	4.35	Buckled	41	351
0.86	1.03	1.60	25.2	58.8	5.1	74,000	81,000	189,200	0.69	6.04	Buckled	41	375
1.07	3.67	1.56	27.9	52.5	5.9	66,000	73,000	185,000	0.51	6.84	7.35	41	364
2.58	3.84	1.62	26.0	39.5	14.3	69,000	90,000	255,000	0.89	1.92	2.81	52	...
2.46	3.67	1.54	30.7	42.4	14.3	88,000	108,000	268,000	0.58	2.15	2.73	49	...

contains very hard carbide spines of the Cr_7C_3 type, which will quickly dull even tungsten-carbide cutting tools. Figures 26 and 27 illustrate how microstructure changes with carbon content.

Abrasion resistance differs considerably. The 2.5% carbon grade has relatively good resistance to wear from abrasive particles like sand, though a number of other alloys will outperform it in such service. In contrast, the 1.0% carbon material is poor for most abrasive conditions. It is quite unsuited for high stress grinding abrasion such as occurs in ball mills.[22, 23] As a principle, selection of these cobalt-base alloys should be based primarily on their properties above 1000° F.

A typically appropriate application is the surfacing of internal combustion engine valves, where resistance to hot gas corrosion and hot deformation is required along with some toughness to withstand the successive impacts of valve action. The 1.0% carbon alloy is widely and successfully used for such work. At room temperature it has a tensile strength of about 100,000 psi. with 1% elongation (as determined on castings) and as temperature rises there is some tendency for ductility to increase.

As carbon increases, room temperature behavior becomes similar to that of cast irons and tensile tests are poorly adapted for comparisons. Since most service conditions involve compressive loading, compression tests are much more appropriate. Relative elasticity, strength and plasticity are compared against carbon (and tungsten) by this criterion in Table 20. Note that the softer and weaker of the two types will deform more readily in compression, but that once plastic behavior begins, with 1.0% C there is a greater reserve of plasticity before failure occurs. These data are useful in judging relative performance in applications where experience has not yet clarified the selection problem. The indicated plasticity is valuable for redistribution of concentrated stresses that might otherwise become dangerous, particularly where some impact is involved.

Some quantitative knowledge of scaling rates that may be involved near the upper useful limit of temperature for these alloys may be required for some applications. For convenience, Fig. 28 has been assembled from cooperative research between the American Brake Shoe Co. and the Alloy Casting Institute. It will serve to supplement Fig. 10. The two materials at the left end of the chart are iron-base heat-resistant alloys of the "HH" and "HK" types; the others show the effect of substituting cobalt for nickel and of adding carbide-forming elements. Tungsten has the least adverse effect of the three shown. Molybdenum should be used cautiously as it may lead to "catastrophic oxidation."[24]

Tungsten Carbide Composites: A.W.S. Class III

Tungsten-carbide is usually exploited for hard facing in the form of granules or small slugs that are intended for welding in place by means of steel rods. The granules are sold in bulk, used for filling steel tubes that serve as composite welding rods, or are otherwise fabricated with one of several binder metals. The tube

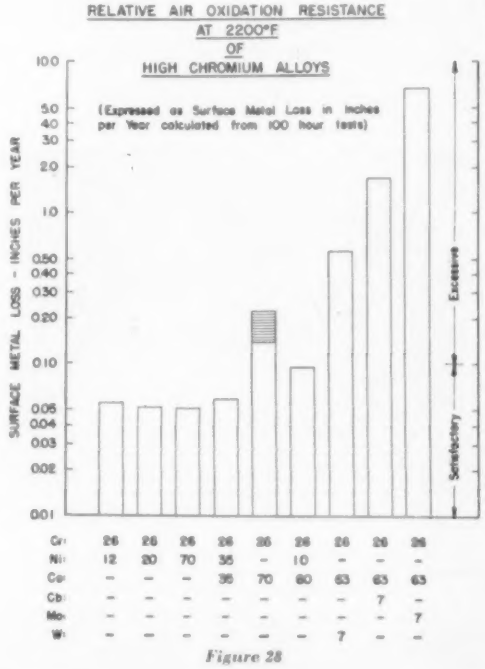


Figure 28

rods are most common. The deposits are used to withstand severe abrasion where a rough weld surface is not objectionable or where a cutting action is required. Surfacing of the rock cutting bits used for oil well drilling is a typical example. Elevated temperatures are not usually involved.

The fused carbide granules probably have high hot hardness but data are not yet available for this survey because the technique for testing them has not been satisfactorily developed. Weld deposits from steel tubes containing these granules are heterogeneous, with structure and properties dependent on several factors.

The granules are sized to conform to manufacturing specifications and may range from around $^1/_8$ in. or more in diameter down to −100 mesh powders. When tube rods are fused the steel from the sheath dissolves some of the tungsten carbide to form a tungsten iron or steel after solidification. Fine granules or powder are dissolved more readily than coarse particles and produce a matrix higher in carbon and richer in tungsten. Similarly, arc welding dissolves more of the granular carbide because of the higher temperatures attained. The matrix composition and character have an important influence on properties of the weld deposit. Some tungsten-carbide is always dissolved by the steel but with a tube rod containing very fine particles and deposited by arc welding, solution may be complete. In this case, the solidified deposit contains none of the original tungsten-carbide but instead is filled with complex tungsten-iron-carbide crystals that form upon solidification, as illustrated in Fig. 29.

The hot hardness of deposits made by both arc and gas welding from tube rods containing fine granules (below 40 mesh) appears in Table 21. The greater amount of tungsten-carbide dissolved during arc welding is reflected in higher hardness at 1200° F. However, this greater solution may adversely affect other properties. Wet quartz sand abrasion factors are included to show the profound effect on one kind of wear resistance.

Table 21—Hot Hardness of Deposits from Composite Tungsten Carbide Rods

A.W.S. Class III-B

Rods are 40% soft steel tube and 60% filler, sized—40 mesh. Individual test specimens were stabilized 24 hr. at test temperature before testing.

Gas weld deposit

Wet quartz sand abrasion factor	Rockwell "C" scale hardness: At room temperature Before	After	Loading interval, min.	Testing temperature, ° F. 1000	1200
0.16	(1000° F.)		0	51	28
	65	59	1	46	3
	(1200° F.)		2	45	− 7
	57	61	4	44	−16

Arc weld deposit

Wet quartz sand abrasion factor	Rockwell "C" scale hardness: At room temperature Before	After	Loading interval, min.	Testing temperature, ° F. 1000	1200
0.67	(1000° F.)		0	50	36
	63	62	1	47	23
	(1200° F.)		2	46	18
	62	62	4	45	13

Arc weld deposit

Wet quartz sand abrasion factor	Rockwell "C" scale hardness: At room temperature Before	After	Loading interval, min.	Testing temperature, ° F. 1000	1200
0.74	(1000° F.)		0	50	40
	65	64	1	46	25
	(1200° F.)		2	46	18
	65	64	4	45	12

Sintered tungsten-carbide powder compacts are customarily brazed to steel shanks or other foundation parts to form cutting tools or other members subjected to extreme wear. This use is obviously related to hard facing by welding. The hot hardness of cemented carbides of this type, as given by Engle,[35] appears in

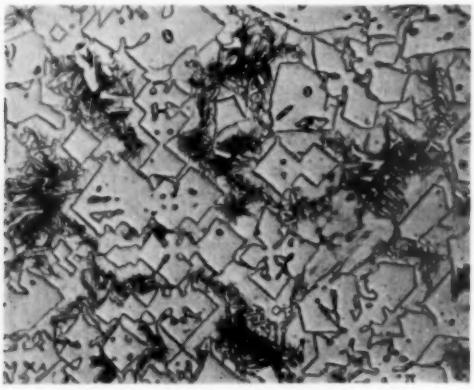


Fig. 29 Photomicrograph at 500 × of an arc weld made from a composite tube rod containing −20/+30 mesh granules of tungsten carbide

The rectangular crystals are secondary tungsten-iron carbides that formed when the weld solidified. The tungsten and carbon were dissolved from the primary carbide by molten low-carbon steel from the sheath, resulting in a new tungsten-iron-carbon alloy. None of the original tungsten carbide granules appear in this picture

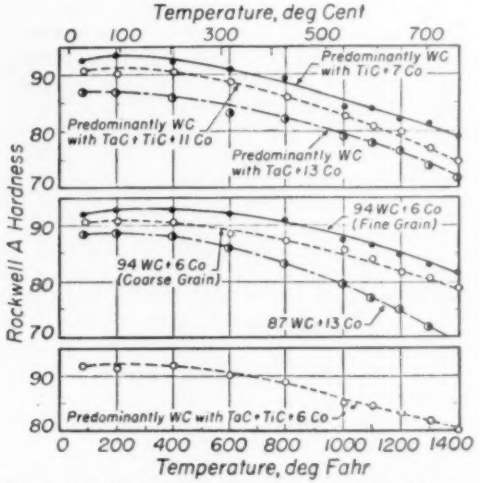


Fig. 30 Hot hardness of cemented tungsten carbide compacts as reported in the 1943 Metals Handbook and by Engle.[35] The testing technique is reported to be similar to that employed in this hard-facing survey

Table 22—Hot Hardness of 18% Cr – 60% Ni Heat-Resistant Cast Iron

A.W.S. Class V-B, Alloy Casting Institute Type HX-250

Chemical analysis, %					
C	*Mn*	*Si*	*Ni*	*Cr*	*N*
2.41	0.77	1.55	59.5	18.0	0.02

One specimen from a small casting tested consecutively at four temperatures without previous heat treatment.

Room temperature hardness: C23 before and C22 after testing.

Loading interval, min.	*Rockwell "C" scale hardness.*			
	600° F.	*800° F.*	*1000° F.*	*1200° F.*
0	17	17	13	6
1	17	17	12	− 8
2	17	17	12	−11
3	16	16	12	−13

Fig. 30. The Rockwell "A" values from 75 to 85 at 1200° F., corresponding to about C48 to C67, are higher than those of the other materials reported here.

Nickel Base Alloys: A.W.S. Class V-A—Nickel-Copper Alloys: A.W.S. Class V-B—Nickel-Chromium Alloys: A.W.S. Class V-C—Nickel-Molybdenum Alloys

Neither copper-base alloys or the nickel-copper (Monel) type have been included in this survey. Nickel-chromium alloys are pertinent, however, since several types have application as hard facings.

A nickel predominating Ni-Cr-Fe alloy, which may be compared with the related alloys in Tables 15–17, appears in Table 22. Its hardness at various temperatures is somewhat lower than that of the chromium predominating alloys.

The addition of tungsten (Table 23) or boron (Table 24) serves to increase hot hardness. The difference between the gas and arc welds of the Ni-Cr-B type suggests that carbon pickup from the oxyacetylene flame may make an important contribution to hardness. These alloys are sometimes described as containing chromium boride in a nickel matrix, though the development of carbides makes the structure more complex.

Sindeband[36] describes a related material, chromium boride bonded with nickel or nickel-chromium alloys. At 1200° F. hot hardness (from the Bishop and Cohen technique) approximates that of some of the sintered tungsten-carbide compacts but 1500° F. stress-rupture fracture times are about one-fourth that of cast vitallium, a cobalt-base alloy similar to the low-carbon Cr-Co-W type but containing molybdenum instead of tungsten.[31] At 1600° F., the stress for fracture in 100 hr. is about 2700 psi., for comparison with 6500 psi. for the HH32 alloy described in Table 11.

Table 23—Hot Hardness of a Ni-Cr-Co-W Alloy

A.W.S. Class V-B

Nominal composition, %				
C	*Cr*	*Ni*	*Co*	*W*
2.5	28	40	9	16

Gas weld deposit tested consecutively at three temperatures without previous heat treatment.

Room temperature hardness: C42 before and C42 after testing.

Loading interval, min.	*Rockwell "C" scale hardness.*		
	800° F.	*1000° F.*	*1200° F.*
0	37	36	31
1	36	35	26
2	36	34	24
3	36	34	23

Table 24—Hot Hardness of Ni-Cr-Boron Weld Deposits

A.W.S. Class V-B

Single specimens tested consecutively at three temperatures without previous heat treatment.

Nominal composition, %		
Ni	*Cr*	*B*
80	11	2.5

Room temperature hardness: C32 before and C31 after testing.

Loading interval, min.	*Rockwell "C" scale hardness of gas weld deposit.*		
	600° F.	*800° F.*	*1000° F.*
0	27	21	17
1	26	18	15
2	26	18	14
3	25	18	13

Nominal composition, %		
Ni	*Cr*	*B*
80	17	3.75

Room temperature hardness: C61 before and C60 after testing.

Loading interval, min.	*Gas weld deposit.*		
	600° F	*800° F.*	*1000° F.*
0	55	52	48
1	54	51	42
2	54	51	41
3	54	50	40

Chemical analysis of a welding rod, %						
C	*Si*	*Ni*	*Cr*	*B*	*Co*	*Fe*
0.67	4.51	71.0	14.8	3.60	0.21	4.85

Room temperature hardness: C53 before and C52 after testing.

Loading interval, min.	*Arc weld deposit.*		
	600° F.	*800° F.*	*1000° F.*
0	49	46	39
1	49	45	33
2	48	45	32
3	48	45	31

Sindeband's air oxidation tests suggest that the formation of low melting point constituents is responsible for the poor hot strength at 1500 and 1600° F. in contrast to the high hardness at 1200° F. He sets 1750° F. as the maximum temperature for usefulness. Note that these are powder metallurgy compacts containing around 70% of CrB. However, the temperature limitation may also apply to welding alloys that have different proportions of CrB.

The nickel-molybdenum alloys, with and without chromium and tungsten, are intended primarily for corrosion resistance.[36–38] However, they also have good hot strength and are sometimes used as heat-resistant alloys. They are usually low in carbon; if gas welded, the corrosion behavior may be changed by carbon pickup from the oxyacetylene flame. Hot-hardness data for small castings appear in Table 25. Note that creep resistance is good but that hardness is low.

Table 25—Hot Hardness of Ni-Mo-Cr-W Alloys

Chemical analysis, %							
C	*Mn*	*Si*	*Ni*	*Cr*	*Mo*	*W*	*Fe*
0.08	0.44	0.50	58.9	15.6	16.7	4.30	3.14

Specimens from small castings stabilized by 24 hr. at the test temperature and tested individually.

Room temperature hardness: C25 before and C28 after testing at 1000° F.; C27 before and C25 after testing at 1200° F.

Loading interval, min.	*Rockwell "C" scale hardness* 1000° F.	1200° F.
0	16	7
1	15	3
2	14	1
4	14	−1

Chemical analysis, %							
C	*Mn*	*Si*	*Ni*	*Cr*	*Mo*	*W*	*Fe*
0.06	0.50	0.42	58.4	11.4	17.2	4.43	7.30

Specimens from small castings stabilized by 24 hr. at the test temperature and tested individually.

Room temperature hardness: C17 before and C16 after testing at 1000° F.; C13 before and C15 after testing at 1200° F.

Loading interval, min.	*Rockwell "C" scale hardness* 1000° F.	1200° F.
0	−1	0
1	−2	−2
2	−3	−3
4	−4	−3

THE SELECTION PROBLEM

Obviously it is not feasible to test all of the available hard-facing alloys with this special technique. In the author's opinion, each important type, or an alloy closely related to it, is included in this assembly of data. Some items have been supplied from experimental melts, or from castings, because they are not well represented as commercial welding rods. Other items in the complete A.W.S. classification system have been omitted because their hot hardness is unimportant; they are useful because of other properties. (The low-carbon stainless steels are an example.) Even this condensation covers too wide a range for convenient selection purposes. If hot hardness were the only property that required consideration, a summary ranking the various alloys would be pertinent; it would be headed by sintered or cast tungsten-carbide, with high carbon cobalt-base alloys, martensitic irons or welds from tungsten-carbide tube rods next, depending on the relative creep resistance needed and the service temperatures involved.

Unfortunately, the selection problem is more complex. Toughness, abrasion resistance, compressive strength and other factors are frequently as important as hot hardness. Toughness can be roughly estimated from carbon content and room temperature hardness, this relation being implied by the subheads in the classification system. However, abrasion resistance is less readily evaluated. Under low stress abrasion[23] carbide hardness and volume are helpful in ranking alloys; under high stress abrasion[5] test data are needed for valid rankings.

With the primary purpose of simplifying the selection problem the detailed data herein have been judiciously narrowed down to nine types, which appear in Table 26. Either averages or values considered typical are given for 1000 and 1200° F. hardness, together with instantaneous and 2-min. loading values to provide indices of creep resistance. Wet sand wear factors representing high stress abrasion are included. Where well-known materials are missing from this condensed summary, the reasons are usually obvious, as in the case of manganese steel embrittlement.

Most of the types in Table 26 have three lines of data;

Table 26—Condensed Summary of Hard-Facing Alloys Useful for Hot-Wear Applications

A.W.S. class	*Type*	*W*	*T*	*Hot hardness, ° F.* 1000 *0*	1000 *2*	1200 *0*	1200 *2*	*Room temperature* Hardness *RC*	*Bhn.*	*W.S.A.F.*
I-A-2	Air hardening, low-alloy steel	G	X	24	14	−17	−50	50	...	0.70
		G	T	13	−9	−22	−71	35	...	0.80
		G	S	36	26	−5	−44	42	340	0.73
I-A-3b	Air hardening, medium-alloy steel	G	X	36	30	11	−11	60	...	0.68
		G	T	15	0	−20	−61	37	...	0.78
		G	S	40	32	8	−26	54	600	0.68
I-A-3c	Martensitic, medium-alloy iron	G	X	42	33	15	−7	60	...	0.32/0.42
		..	S	46	38	19	−8	63	652	0.37
I-A-4d	Martensitic, high-alloy iron	G	X	46	40	33	14	63	...	0.28
										0.38
		..	S	50	44	32	7	62	...	0.33
I-A-5	High-speed steel	G	X	48	45	31	13	63	...	0.52
		G	T	27	16	5	−22	46	...	0.62
		..	S	45	40	23	−3	52	477	0.58
I-C-1	Austenitic, high-alloy iron	G	X	39	31	23	−9	56	...	0.65/0.85
		G	S	40	36	7	1	57	578	0.67
II-A	Chrome-cobalt-tungsten, No. 6 type	G/A	..	30	29	24	22	43	...	1.10/1.30
		G	S	31	30	24	20	44	387	1.17
II-B	Chrome-cobalt-tungsten, No. 1 type	G/A	..	44	44	39	33	55	...	0.50/0.75
		G	S	45	44	40	34	54	...	0.74

NOTES: W = weld method: G = gas; A = arc. T = treatment before hot testing: X = none; T = tempered 24 hr. at 1200° F.; S = stabilized 24 hr. at test temperature. 0 and 2 indicate loading interval in minutes. W.S.A.F. = wet sand abrasion factor vs. 1.00 for S.A.E. 1020 steel. Hardness numbers are Rockwell "C" scale values.

the first was obtained from welds tested without further heating, the second from welds after tempering 24 hr. at 1200° F. and the third from specimens stabilized 24 hr. at the test temperature before testing. The third lines cover tests made especially for inclusion in this summary and thus have the status of check determinations. The associated abrasion factor was obtained on the same weld deposit. The abrasion factors on the first two lines represent the range usually expected from this type.

The inclusion of data for the as-welded condition without an associated test after 1200° F. tempering indicates that the composition is relatively indifferent to such tempering.

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The valuable assistance of L. S. Gordon, C. R. Wilks, W. S. Pellini,* H. J. Chapin, A. Z. Mather, and other members of the Metallurgical Department is acknowledged with appreciation. The author is also indebted to the American Brake Shoe Co. for permission to publish this material.

SUMMARY

In this survey of hardness over the range from room temperature to 1200° F., two observations are noteworthy: the very great loss of hardness that can occur as temperature is raised, and the marked influence of time on apparent Rockwell hardness.

The quantitative data have been presented in many tables and charts in an attempt to cover industrially important or otherwise significant hard-surfacing alloys that have been commercially available for a reasonable time.

For this wide range of materials one hardness ranking is not appropriate, as the alloys respond differently to temperature. At 600° F. and possibly at 800° F., the room temperature hardness ranking is reasonably valid and the time effect, as evidenced by creep, is not great. At 1000° F., those alloys that exhibit marked secondary hardening tend to show superiority. This advantage is reinforced by carbide volume, as exemplified by the chromium-tungsten martensitic iron type.

At 1200° F., creep is prominent and all of the iron-base alloys show marked softening. Chromium-nickel or chromium-cobalt matrix compositions begin to demonstrate superiority, the latter being stronger and harder. This superiority is expected to be maintained up to the maximum temperature of usefulness, which is probably between 1900 and 2150° F. Carbide volume and tungsten also influence the cobalt-base alloys. Two well-known grades permit a choice between 1% carbon with good toughness but lower hardness and 2.5% carbon with high compressive strength and hot hardness associated with greater brittleness.

Tungsten-carbide inserts welded in place probably provide the highest available hot hardness. Weld deposits from composite tungsten-carbide rods may give deposits with excellent hot hardness, but these are likely to be sensitive to welding technique. Arc welds from these rods tend to result in higher hot hardness but lower abrasion resistance than gas welds.

Gas welds are generally to be preferred for maximum hot hardness because of maximum carbon pickup and minimum dilution from base metal fusion. The composite tungsten-carbide rods are an exception.

Several factors in addition to hot hardness should be considered in the selection of materials for hot wear service. Table 26 was prepared to facilitate and simplify this selection.

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Slope Control and Its Effect on Spot Welding

Discussion by J. L. Solomon

Mr. Johnson has given in his paper a method of solving one of the problems associated with the welding of aluminum. In writing of his findings, and in his experimentations with the slope control, he is verifying the work done at Sciaky Bros., and at numerous Aircraft and other metal-working plants doing production welding of aluminum to ANW-30 specifications and also commercial standards.

When the Sciaky three-phase welding system, the main circuit of which is shown in Fig. 1,[1] was first developed, it was determined at that time that the tip pickup and the resultant tip cleaning problem was reduced considerably from that experienced when welding aluminum with single-phase or storage of energy type of equipment. It was shown that the slow rise in secondary current, Fig. 2, which is obtained naturally in the secondary of a three-phase welder, was the primary reason for this reduction in electrode pickup and increased tip life when welding aluminum alloys. The addition of slope control to a single-phase welder allows one to simulate the heating effect produced naturally by the three-phase type of welding current, and just how close the slope control simulates the heating effect of the three-phase welder can be seen from Fig. 2. The welding of aluminum to ANW-30 specifications, however, requires more than just the solution of the tip pickup problem. The ANW specification requires certain standards as regards the structure of the weld, porosity, sheet separation and internal cracks in the weld nugget. There are other practical problems encountered in the welding of structural parts which make it necessary to give much thought to the prevention of warpage during the welding operation, and,

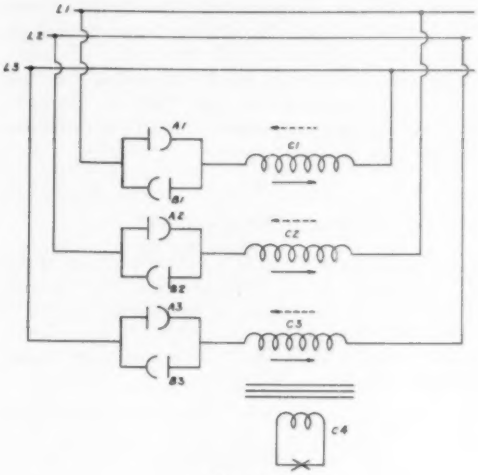

Figure 1

of course, because of the high currents required in the welding of aluminum and aluminum alloys, there always exists the problem of finding an adequate power supply to operate the welding machine. The addition of slope control to a single-phase welder does not solve any of these latter problems. The three-phase welder was designed originally with the thought in mind of solving as many of the above problems as possible, and the success obtained in solving these problems is verified by the many machines installed in Aircraft and

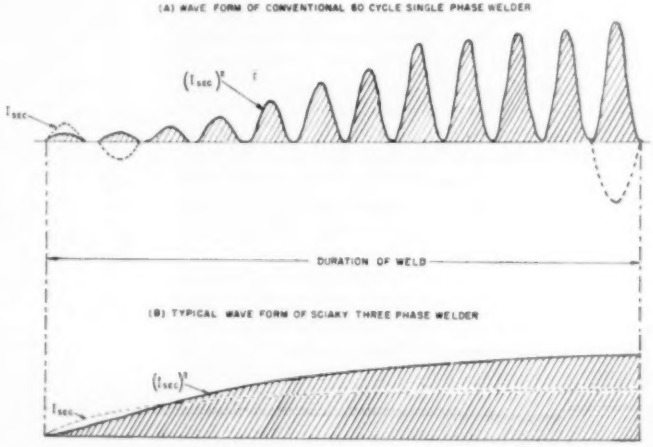

Fig. 2 Relation between wave form and joule effect

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The paper by Ivar W. Johnson was published in the October Welding Research Supplement, pp. 471-s to 476-s.

[1] Solomon, J. L., "Three-Phase Balanced Load Resistance Welding Machines," THE WELDING JOURNAL, **26** (5), 426–431 (1947).

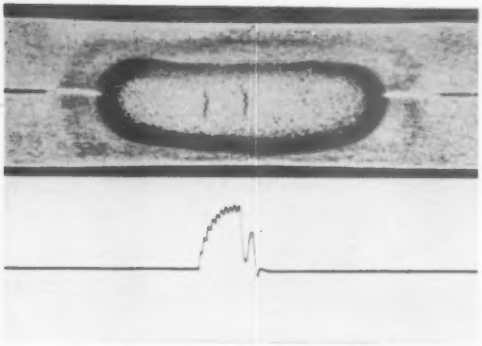


Fig. 3 Welded without current decay control (0.064 + 0.064 24ST Alc.)

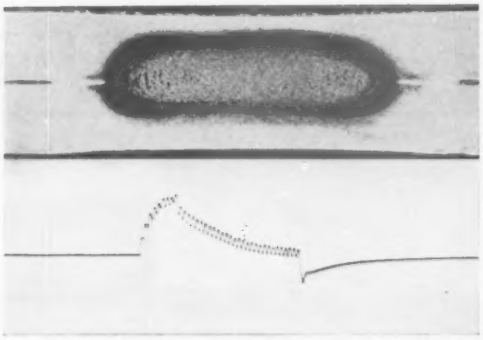


Fig. 4 Welded with current decay control (0.064 + 0.064 24ST Alc.)

other plants which require the welding of aluminum to ANW specifications.

In order to produce a weld which is free from porosity, it is necessary to apply adequate force during the welding operation and sufficient recompression following the weld period to forge the metal. The use of low forces during the welding and forging operations results in a weld nugget which is spongy and porous.

Whenever long welding times are utilized in the welding of aluminum, the problem of warpage creeps in and this is indeed a serious problem when one considers first, the necessity of maintaining the predetermined contour of airfoil sections for purely aerodynamic reasons; and secondly, that subassemblies which are being spot welded must be fitted together on one general assembly. Warpage of any part makes this fitting a very difficult operation.

After the weld nugget is formed in heat-treatable aluminum alloys, shrinkage cracks will result if the weld nugget cools too rapidly. Figure 3 is a cross section of a weld made in 0.064 plus 0.064-in. 24ST Alclad, utilizing a welding current with slowly rising current which terminates abruptly. The vertical lines at the center of the weld nugget are shrinkage cracks caused by too rapid cooling of the weld nugget.

Figure 4 is a micrograph of a weld made in the same material, using the same welding current but with the addition of a current decay control which has the function of allowing the spot to cool at a slow rate. The wave form of the secondary current used in producing this weld is shown in Fig. 4.

In order to prevent warpage of the sheet, a short welding time is required. For example, in the welding of 0.064–0.064-in. 24ST aluminum, a weld time of 3 cycles with current decay time of 9 cycles—was utilized. The force used during the welding was 1000 lb.—forging force, 2900 lb.—peak secondary current, 49,500 amp.

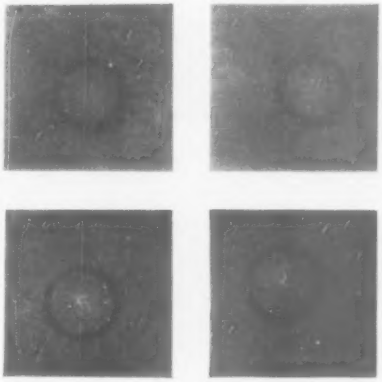


Fig. 5 (Top) Welded with current decay control; (bottom) welded without current decay control. (0.064 + 0.064 ST Alc.)

To indicate the power demands required to properly weld this material on a three-phase machine as against a single-phase machine, it must be noted that the three-phase power demand for a 36-in. throat machine is 192 kva., whereas the power demand for the single-phase machine with the same throat depth would be 760 kva.

Although it is possible to make welds in heat-treated aluminum which are acceptable for ordinary commercial use, it is difficult to set up a welding machine with or without slope control so as to consistently obtain welds which will comply with ANW-30 specifications in all respects. The addition of the current decay control to the three-phase welding current pulse makes it possible to easily set up a welding schedule and easily weld aluminum alloys to Air Force Navy specifications.

PRACTICAL WELDER AND DESIGNER

Fig. 1 Resistance welded hot air heating chamber

Resistance Welding Hot Air Heating Chamber

by H. W. Stieglitz

YOU are looking at the hot air heating chamber of a gas furnace. It consists of three closely spaced chamber sections linked by a number of tubular flues. The entire assembly must be gastight. Problem: How to go about welding the flanged tubes to the chamber sections at high-production speeds, keeping in mind the limited space between sections?

The solution to this interesting problem involves two seam welders, one of which is an entirely new, patent-applied-for application of seam welding. Once this new machine was developed for welding the flues to half sections of the chambers, it became a simple matter to complete the assembly by welding the chamber halves together with a standard circular seam welder job adapted to facilitate that part of the work.

H. W. Stieglitz is Chief Engineer of the Thomson Electric Welder Co.

The key to the speed and simplicity with which this difficult assembly operation is performed, is the special 150 kva. horizontal seam welder shown in Fig. 2. It consists essentially of a pair of horizontally opposed, motor-driven rotating heads with a stationary, air operated, gripper arm between. The rotating heads are backed up by air cylinders to provide welding force.

The work procedure is to preassemble a set of flue pieces and the halves of the two chamber sections which they connect; as for example, the three flues, the upper half of the bottom section and the lower center section which are bracketed in Fig. 3 as "Step 1." Assembly is facilitated by vertical flanges around the holes in the chamber halves which are raised for this purpose when these parts are formed. The weld is made between the $^1/_2$-in. flanges on the flues and the surface of the chamber section.

Fig. 2 Production starts here—with the work held stationary in the gripper clamp, the flues are quickly welded to adjacent halves of the chamber sections

Both ends of each flue are welded at once except for two flues which require only one weld each

This subassembly is placed in the welder in a vertical plane so that the jaws of the gripper arm surround one of the flues. The gripper is clamped by tripping one of the twin-foot switches. The other foot switch is a two-stage switch which first brings the welding heads into contact with the work, applying welding

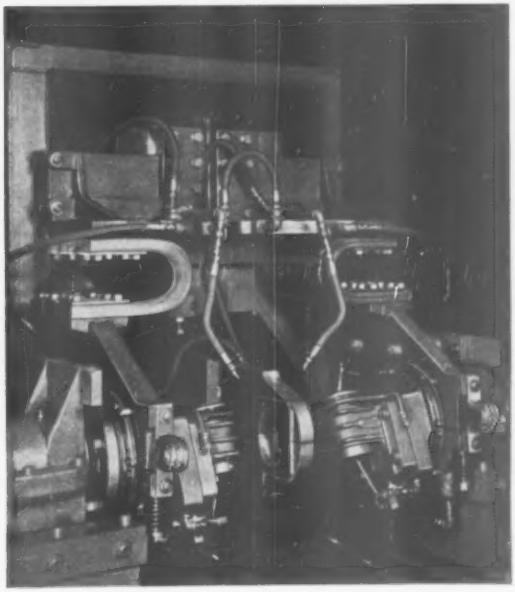


Fig. 4 This close-up shows details of the gripper clamp and how the revolving heads are connected to the secondary transformer leads

Note the provision for complete and effective water cooling

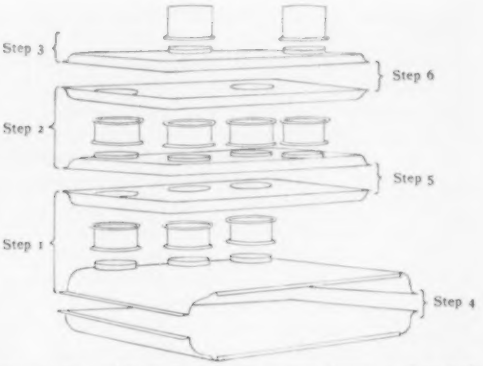


Fig. 3 By welding the chambers last (steps 4, 5 and 6) the complete heating chamber is easily assembled

Steps 1, 2 and 3 are performed in the special purpose welder

force and then initiates rotation of the heads.

Two time delay relays are used: one to delay application of welding current until head rotation is up to speed, and one to cut out the entire circuit so that welding current is cut off and the heads withdrawn as soon as the weld is complete. There is no need to drive the welding wheels since contact with the work insures uniform rotation. A seam welding timer interrupting

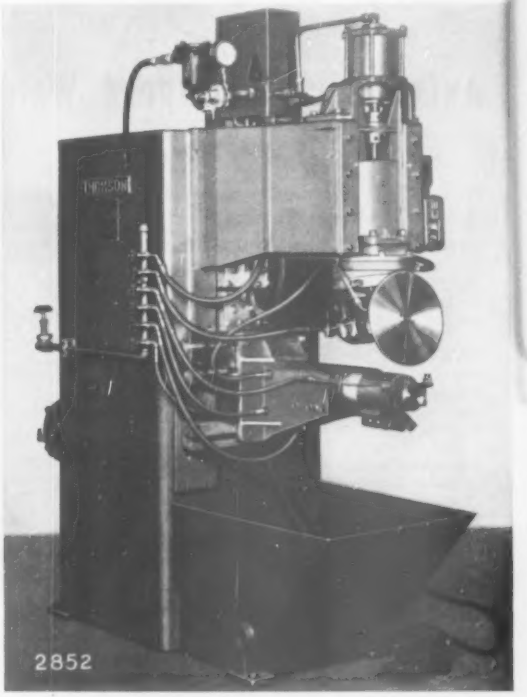


Fig. 5 Tilted upper wheel of this circular seam welder helps keep the weld close to the inner edge of the flange

Small lower wheel is designed to go between chamber sections. Guides insure straight seams, easy corner turning

the primary power establishes the required gastight seam-weld pattern.

This process is repeated to weld each flue in the first subassembly and continued through steps 2 and 3. You will note that in steps 1 and 2, two welds are made simultaneously to join the two chamber halves to opposite ends of each flue. In step 3, only one chamber half is involved but the procedure is exactly the same. The head which is not welding simply provides back-up pressure to prevent deflection of the gripper arm and electrical connection to the secondary leads which, as can be seen, are connected directly to the welding heads.

The final assembly is a routine procedure in which mating halves of the chamber sections are rim welded by a 150 kva., circular seam welder (Fig. 4). This standard welder is slightly modified to suit the nature of the job. Notice the unusually small lower wheel which is designed to fit readily between chamber sections as the assembly is built up. Notice, also, the guides at either side of the lower wheel which facilitate running straight, close seams along the sides and easy corner turning.

The upper wheel and its mounting are also job adapted to permit welding as close as possible to the inner edge of the flange. The 8° tilt makes a big difference in work handling but in no way impairs the solid application of welding force or the even passage of welding current. The drive is through the lower wheel, with the upper wheel idling.

This job is also an interesting demonstration of the application of special purpose and standard welders in their respective fields. The special purpose machine does a job which could not be accomplished in any other way. While it involves considerable engineering, it is obvious that it handles the job in a simple and efficient way and that the cost-per-weld is low. The standard machine, on the other hand, involves a minimum of special tooling. It, too, does the job required in the simplest and least expensive way.

Automatic High-Speed Welding of Small-Diameter Joints

TO ATTAIN high-production speeds yet sound joints in the critical welding of universal yokes to each end of a $2^3/_4$-in. diameter automobile shaft, the Mechanics Universal Joint, Division of Borg Warner in Memphis, Tenn., has installed a completely automatic production welding machine.

The shafts are fed to the machine on a conveyor and can be seen in the picture entering from the left. Yokes are brought to the machine on a conveyor from the right. Both yokes and shafts have a small chamfer to produce a vee joint.

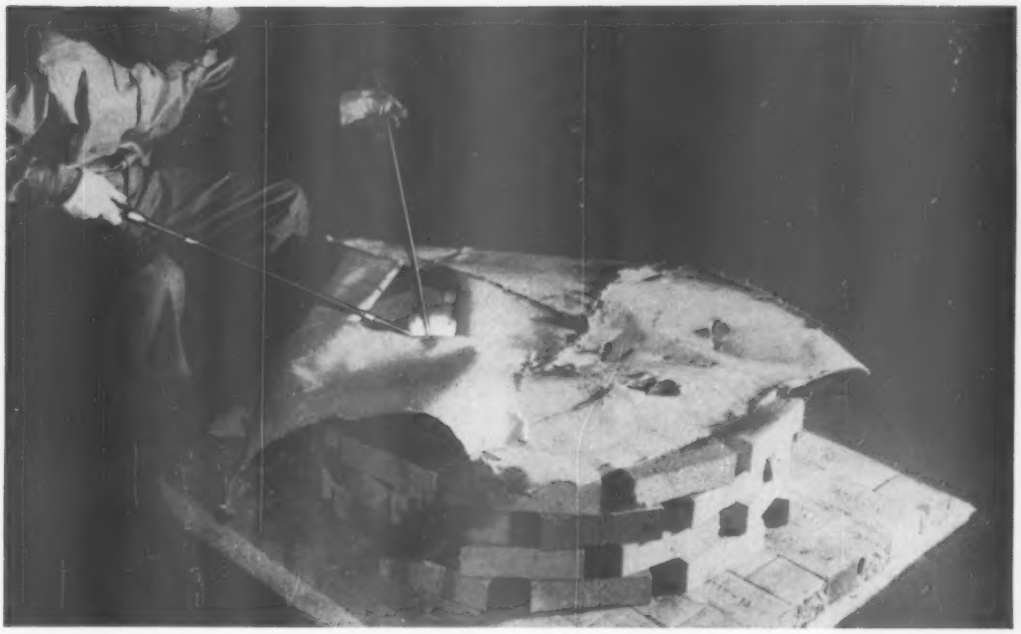
The operator hydraulically clamps the yokes into the shaft in the small jig in front of him and then places the assembly in one of the four positions in the turret head of the production unit. He starts the operation which then continues automatically, requiring only the placing of the work in the machine.

The turret rotates putting the shaft in the welding position under the two welding heads. As soon as the operator places another shaft in the front position of the turret, the wire electrode starts to feed down, the flux valves open, the arc starts and two welds are made simultaneously. A small piece of fiber $1^1/_2$ in. square is attached to the head to make a dam to prevent the flux spilling off the shaft. The two welding heads are standard automatic units employing the hidden arc process of welding.

Photo and data courtesy Lincoln Electric Co.

The welds are made in 11 sec., using $^3/_{32}$-in. electrode wire, 325 amp. and 26 v. When the weld is completed, the flux valves close automatically, the arc and wire feed stop and the turret revolves positioning another shaft for welding and discharging a completed assembly at the rear of the machine.

The machine was designed and made by the Cecil C. Peck Company of Cleveland, Ohio.



How to Fusion Weld Cast Iron

by F. C. Geibig

THERE are three types of cast iron: white cast iron—used for plowshares, car wheels, stamp shoes, dies, wearing plates—which is seldom repaired by welding shops; malleable iron—used for hardware and small tools, household appliances, farm implements, heavy machine parts—which can be repaired only by braze welding; gray cast iron—used for cook stoves, automobile cylinder blocks, hydrants, machine bases and hundreds of other things—which can be repaired by braze welding or fusion welding.

Table 1 helps you identify the cast irons and malleable iron. But, if you cannot identify the part, repair it by braze welding—it may be malleable iron, and the heat of fusion welding would destroy its malleability.

Chances are that most cast-iron parts you have to repair will be gray cast iron. And if the parts are used at temperatures above 500° F., or if the weld must match the color of the casting, you must fusion weld with cast-iron rod. This article tells you how to fusion weld gray cast-iron parts.

F. C. Geibig is connected with The Linde Air Products Co., Newark, N. J.

HOW TO WELD SMALL PARTS

The first step in a repair is to prepare the parts for welding. Usually the broken edges should be beveled. However, if you bevel small parts, like the one shown in Fig. 1, it will be almost impossible to make an exact fit. With a small part like this it is easier to melt out a vee with the blowpipe flame after tack welding.

Table 1

White cast iron	*Gray cast iron*	*Malleable iron*
Fracture shows very fine *white* silky particles. Very brittle. When chipped, breaks into very small fragments	Fracture shows tiny *gray* flaky particles. Broken chips about $\frac{1}{8}$ in. break off from base metal	Fracture shows fine round *gray* particles. Very tough. Hard to chip. Broken chips will curl up and be longer than gray cast iron

LINE UP THE PARTS

The next step is to line up the parts. Prop the larger piece between bricks to keep the part steady. Now balance the broken end on the original break as shown in Fig. 2. This will give you an exact fit. Sometimes parts can be lined up with clamps, in a vise, in sand or with carbon paste.

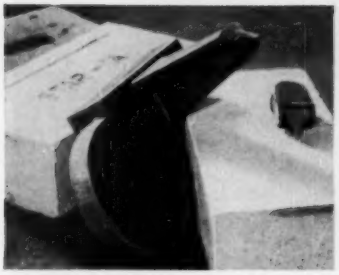


Fig. 1 You can set up small parts with bricks. Use small pieces of scrap metal, too, to get the break about level

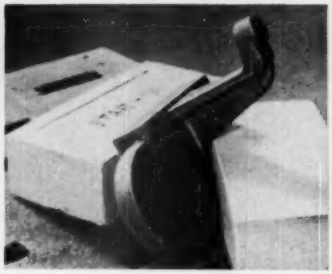


Fig. 2 When the large section is held securely by the bricks, balance the small one on the break

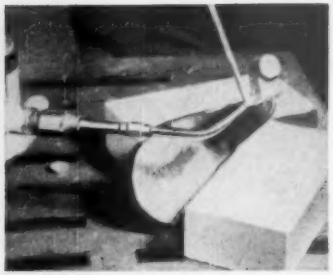


Fig. 3 Make tack welds at each end of the break. They should be strong enough to hold the parts during welding

MAKE TACK WELDS

After the parts are lined up several tack welds should be made to hold the parts in place. Attach a small size welding tip to the blowpipe and adjust the flame to neutral.

Now heat the part and play the flame all around the break. At the same time heat up a cast-iron welding rod and dip the hot end in cast-iron welding flux. Use $^1/_8$-in. rod or smaller. Concentrate the flame on a spot at the break and melt a small puddle. Add molten welding rod to the puddle and build up a small weld as shown in Fig. 3. Make another tack weld on the other side and let the welds cool.

SUPPORT PARTS WELL AND PREHEAT

Next you have to get a good setup and preheat the part for welding. Place the part so that you can weld on the horizontal. Use the bricks again as a support. For larger castings, good support is more important since cast iron tends to sag when it is heated and may warp from its own weight.

MELT OUT A VEE

Warm the entire casting, then concentrate the flame along the crack until the metal turns red. Then melt out a vee along the crack. The crack will appear white through welding goggles. Heat along this white line and use a welding rod to scrape away the molten metal to form a vee about $^1/_8$ in. deep. Be careful not to melt out the tack welds.

BEGIN WELDING

Now you are ready to begin welding. A vee has been melted and the part preheated at the same time. Move the blowpipe to one end of the vee and play the flame on one spot. At the same time, bring the rod into the flame.

When metal from along the vee melts and forms a little puddle, dip the rod into the flux and then place it in the puddle. The end of the rod will melt and mix with the puddle. Move the puddle along the vee, melting more metal from the sides of the vee and from the rod.

Be sure to melt the sides of the vee ahead of the advancing puddle. When you add rod metal do not let it drip into the puddle. Keep the end of the rod in the puddle and move it around gently to work out gas bubbles.

If a little metal runs over the end of the vee it will form sort of a shelf and make it easier to deposit more weld metal. Use a good flux to float out impurities. Skim specks and dirt off the puddle with the end of the rod.

Build up the weld so that it is slightly higher and wider than the base metal. After you finish welding this side, let the weld cool. Then turn the part over so you can get at the web and sides of the part. Again adjust the bricks and make the part steady. Begin melting the web of the part at the crack. Melt this metal and scrape it away with a rod until the white crack in the base metal disappears. When the white line disappears, you are down to the weld metal just deposited.

Now start at one end of the vee and form a puddle again. Add rod metal and build up the weld higher than the base metal, just as you did on the other side. Be sure to build up the web you melted away, too.

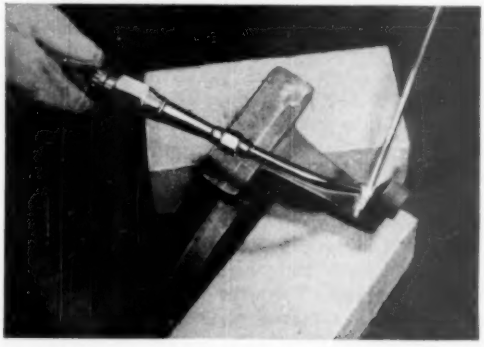


Fig. 4 Make a vee across the break by melting the metal around the crack and then scraping it away with a rod

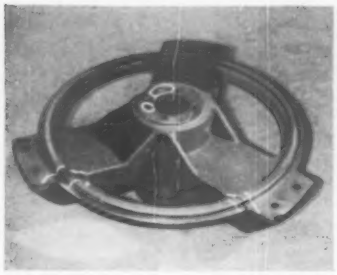


Fig. 5 This casting was cracked between webs at the rim. There were also defects in the hub

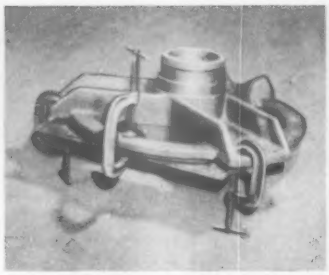


Fig. 6 Then line up the broken rim with the original breaks so that it rests on the steel bar. Vees are face down

Fig. 7 Make tack welds at each break and when they cool, remove the clamps and steel bar. Then build up the furnace

Next, weld the casting at each end where you made the tack welds. Melt and scrape away the tack welds until you are down to the weld metal. Then weld these sides. Cover the part with asbestos paper or bury it in slaked lime so that it cools slowly.

When the part is completely cool, remove the excess weld metal on a grinder. The part will then be returned to its proper shape and the weld will match exactly the color of the original casting.

HOW TO WELD LARGE PARTS

To fusion weld a large casting, like the one shown in Fig. 5, exact line-up, good support and proper preheat are especially important. Preheating is important because it equalizes the stresses that might crack the casting during welding or while cooling. It also relieves stresses that may exist in the casting.

The first step is to bevel the small rim section on a grinder. Make a single vee at each end of the section, but leave about $^1/_8$ in. of the original break at the bottom so that the parts line up exactly.

Now chip out a bevel at each end of the main piece to match the vee of the smaller section. Use a cold chisel and hammer. Leave about $^1/_8$ in. of the original break at the bottom at each end. This untouched section will help get an exact match. Then use the chisel and hammer to vee the cracks in the hub.

Fig. 8 Build up the raised section slightly higher than it was originally. Melt the sides of the vee and deposit rod

MATCH PARTS EXACTLY

Place the main part of the casting so that the bevels are facing down as shown in Fig. 6. Then place straight steel bar, about $^1/_2$ x 4 x 18 in., across the gap so that it rests on machined surfaces of the casting. Use C-clamps to hold this bar to the casting at each end.

Line up the rim section with the larger part of the casting. It should rest on the steel bar. Check the original break at each end of the broken rim to line it up exactly. This may require light tapping with a hammer. When the small section lines up with the big section, clamp it to the bar with a C-clamp.

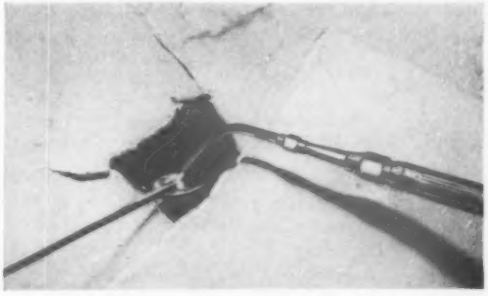


Fig. 9 Melt the sides and bottom of the defects, then add welding rod. Build up higher than the rest of the hub

Fig. 10 After machining, the finished part looks like this. The perfect color match makes it difficult to see the welds

Fig. 11 Pick a spot in your shop that has a flat floor. Place asbestos paper over the floor so that it covers an area about 4½ x 4½ ft.

Fig. 12 Then lay fire brick to make the furnace floor. Stagger the bricks just like bricklayers do, so that one row laps the end of the other

TACK WELDS HOLD PIECES

Attach a medium size welding tip to the blowpipe and adjust the flame to neutral. Make tack welds at each break. First heat the entire casting slightly with the blowpipe flame. Then concentrate the heat on a spot at the outer edge of the rim as shown in Fig. 7. At the same time, heat the end of a cast-iron welding rod and dip it into the flux. When a little metal melts and forms a molten puddle, add molten rod and build up a weld about the size of a dime. Then let the tack weld cool. The tack weld should be about ⅜ in. thick at its center. Make a tack weld at the other break the same way.

BUILD A FURNACE

Now build a furnace, like the one shown in Figs. 11 to 17, to heat the part evenly. Follow the instructions carefully. First make the furnace floor, then place the casting in the center of the floor. Next remove the clamps and steel bar and turn the casting over so that the vees are on the top side. Be sure to support the casting as shown in Fig. 13. Mark the casting with a 600–800° F. temperature-indicating crayon. Then add charcoal, light it and cover the furnace.

PROPER PREHEAT IMPORTANT

A casting this size needs plenty of heat—about 700–1000° F.—so you have to preheat it carefully. If you try to preheat the casting with the blowpipe, you will set up stresses that might crack it. A charcoal furnace will heat the casting evenly. It is also more economical to use charcoal rather than oxygen and acetylene to bring the casting up to welding temperature. It is a good idea to use a temperature-indicating crayon for a large preheating job like this. Then you will know when the casting is ready to be welded.

WELD CASTING IN FURNACE

When the mark melts, you are ready to weld. Remove part of the asbestos roof so that you can get at the cracks. Keep the rest of the casting covered. Heat the vees and surrounding areas with the blowpipe flame,

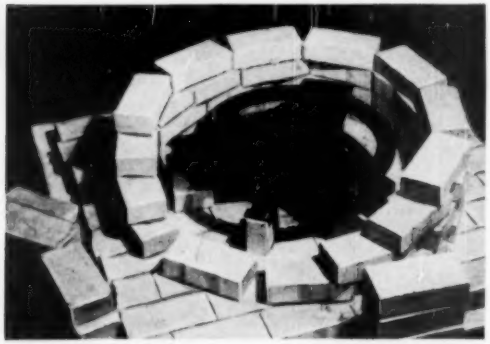


Fig. 13 Now place the casting in the center of the floor. Leave about ½ in. space between the bricks in the first course. Lay the other courses so that the ends are together. Support the casting at three different places

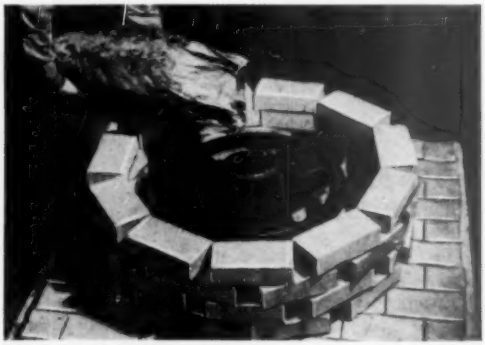


Fig. 14 Line the floor with charcoal. Keep it away from around the hub to prevent the hub from getting too hot and expanding. Do not use coal or coke since they do not burn as uniformly as charcoal does

Fig. 15 Light the charcoal with the blowpipe flame. Do not use kerosene or gasoline to start the fire

Fig. 16 When the fire starts, cover the top of the furnace with asbestos paper. Two bars act as supports

Fig. 17 Put vent holes in the asbestos paper. After welding let the casting cool in the furnace

then start welding. Hold the blowpipe right over one of the vees, and move the flame in small circles.

Now concentrate the flame at one end of a vee. Heat both sides evenly, with the inner cone about 1/4 in. from the metal. Heat the end of a cast-iron welding rod at the same time. Then dip it into the flux can.

Keep playing the flame over the part until the bottom and sides of the vee start to melt and a puddle forms. Add rod metal to the puddle, then move the puddle along the vee just like you did on the small part.

Be sure to melt the sides and bottom of the vee before adding rod to the weld. If rod metal is added before the base metal is melted, the two will not mix together. Use a good flux to float out impurities. Work out gas bubbles, and skim off any specks of dirt again with the end of the rod. Build up the entire weld sightly higher and wider than the original casting.

Just before the puddle reaches the end of the rim, lift the blowpipe away. Then carefully melt the metal along the sides and add just a little rod. Do not form too large a puddle. A large puddle will run off the edge.

After you finish welding one of the vees, weld the other one the same way. Then weld the cracks in the hub. Build up both cracks so that the weld metal is slightly higher than the base metal.

TURN CASTING OVER TO FINISH WELD

After you build up the hub you are almost finished. Remove the asbestos paper roof and supports. Then turn the casting over. Use tongs, a crane or a hoist, whichever is handiest. Make the furnace roof again, and turn back part of the paper to get at the cracks.

Now play the blowpipe flame over one of the tack welds you made and melt it. Use a length of welding rod to scrape away the metal when it melts just like you did before on the small part. Heat the entire crack, and scrape the tack-weld metal and base metal away with the rod. Keep scraping the metal away until the white line in the base metal disappears. When you are down to the weld metal, you are ready to weld.

Continue along until you complete the weld across the entire rim as before. Deposit enough rod, just as you did before, so that the weld is built up slightly higher than the casting. Make the weld at the other break the same way.

SLOW COOLING IMPORTANT

After welding cover the furnace roof completely and let the fire burn out. Leave the casting in the furnace until it is completely cold.

It is very important to cool the part slowly. Sudden cooling of a cast-iron part will produce hard spots in the weld area. This area will be very brittle and so hard that machining might be impossible.

After the casting is completely cold it is ready for finishing. Remove excess weld metal and grind or machine to the proper shape. The finished casting is shown in Fig. 10. A perfect color match makes it difficult to see the welds.

POINTS TO REMEMBER

1. Place the rod in the puddle when you add weld metal—do not let it drip off.
2. Melt and fuse the sides of the vee before adding rod to the weld.
3. Use a good flux to float out impurities.
4. Work out all gas bubbles. They form blowholes and pockets that make a weak weld.
5. Skim specks and dirt off the puddle with the end of the rod.

Biographical Sketches—Nominees for Offices Selected by Nominating Committee

(*Continued from June issue The Welding Journal*)

DISTRICT VICE-PRESIDENTS

District No. 4—Thomas J. Crawford

Mr. Crawford, a native of Pittsburgh, was educated as a Mechanical Engineer at Harvard Engineering School, Class of 1925.

After a year with Youngstown Sheet & Tube Co. at Youngstown, Ohio, as an Industrial Engineer, he joined Truscon Steel Co., where he served until 1931 in various capacities as designer, field sales engineer and department head. Much of the work here involved the engineering and development of welded structures.

From 1931 until early 1934, he collaborated with G. J. Nord, Consulting Textile Engineer, of Gastonia and Asheville, N. C., in the design and development of various textile mills and processes.

In 1934 he joined Republic Steel Corp. at Youngstown. There he set up and headed the Operating Development Dept. with the title of District Research Engineer. The original purpose of this department was to provide the technical organization for the development of the electric resistance method for production of welded pipe. After several years, the scope of the department was extended to cover all of the research and development activity of a completely integrated steel plant.

In late 1944, he left Republic to enter consulting practice in the welding field under his own name. In 1946 he moved from Youngstown to the Detroit area. His activities in this field, as consultant for many well-known firms, have embraced most of the areas of welding processes and design for welding.

He has addressed many A.W.S. Sections on welding subjects and has delivered several papers at Annual Meetings. For the past three years he has been an officer of the Detroit Section, retiring as Chairman this year.

District No. 6—A. E. Wisler

A. E. Wisler is a native of the State of Kansas and attended high school there. He received his B.S. Degree in chemistry in 1935 at the University of Arkansas. Mr. Wisler went to work for Hughes Tool Co. in 1935 in the Chemical Laboratory. After a year in the Chemical Laboratory he began working on research and development problems connected with welding and hard facing. This developed into the present position in the Metallurgical Department, of Welding Engineer.

He is a member of the AMERICAN WELDING SOCIETY, American Society for Metals, Alpha Chi Sigma (Professional Chemical Fraternity) and the Houston Engineers Club, as well as being a registered professional engineer in Texas. He has served on various committees and in minor offices of the other societies but has been most active in the AMERICAN WELDING SOCIETY. In addition to serving on the various committees he has been Secretary-Treasurer for several years and was Chairman of the Houston Section during 1948–1949.

DIRECTORS-AT-LARGE

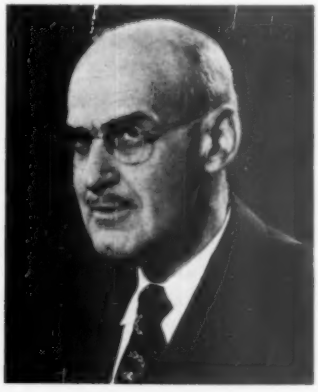
La Motte Grover

Mr. Grover has been associated with Air Reduction during the last eleven years, assisting designers, fabricators, railway car shops, constructors and shipbuilders in the development of design and construction practices for the building of welded bridges, railway cars, buildings, ships and other engineering structures and equipment.

He has been intimately associated with various welding research activities directed toward the development of suitable design practices, materials and construction and fabrication procedures. He is Chairman of the Structural Steel Research Committee of the Welding Research

Thomas J. Crawford

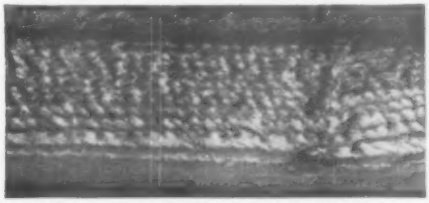
A. E. Wisler



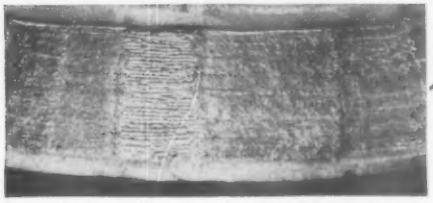
La Motte Grover

BLAST FURNACE UNIT SHOWS NO WEAR

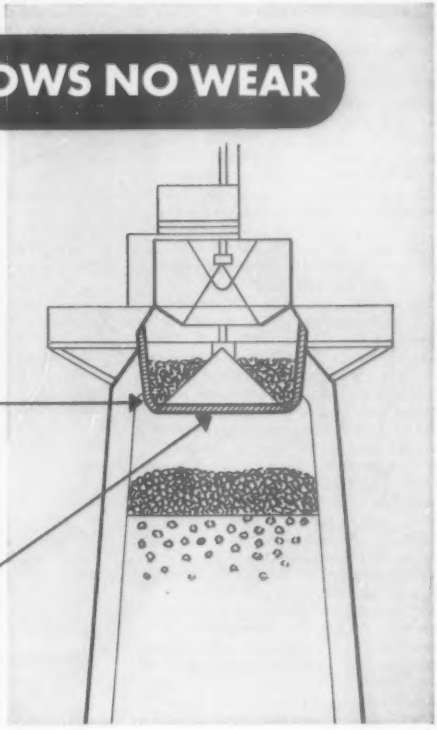
because both BELL and HOPPER were hard-faced with "HASTELLOY" alloy C



Close-up view of hard-faced surface of hopper.



Section of HASTELLOY alloy C deposit on bell.



After a year's service in a top-pressure blast furnace, the HASTELLOY alloy C surfaces on this bell and hopper unit show no signs of pitting, erosion, or wear. The unit resists the abrasive action of various charge materials and the effects of the 250 to 400 deg. F. top temperature.

HASTELLOY alloy C stays hard at the operating temperatures, and deposits made with this alloy maintain their resistance to checking and chipping. Surface hardness can be increased to as much as 400 Brinell by simple cold-working.

Alloy C has already increased the life of other types of steel plant equipment 10 times or more in such applications as blooming-mill hot shear blades, entry guides, twist guide rolls, angular-type guides, and dies and shafts. In these applications, as with the bell and hopper unit, hard-facing results in fewer shutdowns, less replacements, and greatly reduced maintenance costs.

Write for the booklet, "New Hard-Facing Techniques for Protecting Steel Mill Equipment from Wear." It describes some of the many hard-facing jobs that have resulted in large savings through use of HASTELLOY alloy C. In addition to HASTELLOY alloy C, there are many other HAYNES hard-facing rods—for practically every condition of wear, heat, impact, or severe abrasion. For information about any particular application that you may have in mind, contact the nearest Haynes Stellite District Office. Our engineers will be glad to help you.

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Council and a Chairman of a number of other committees and subcommittees of the AMERICAN WELDING SOCIETY and the Welding Research Council, on fatigue testing and impact testing of welded structural details, and specifications for design, workmanship, welding procedures, etc.

He is the author of a *Manual of Design for Arc Welded Steel Structures* published by Air Reduction, and author of many technical articles on welded steel fabrication and construction which have appeared in European magazines as well as technical publications in the United States; also has given technical lectures in all parts of the country on these subjects.

He is a Member of the American Society of Civil Engineers and a member of the American Society for Testing Materials, the AMERICAN WELDING SOCIETY, etc., and is a registered professional engineer. During a period of twenty years prior to joining the staff of Air Reduction, Mr. Grover held various engineering positions, serving some eight years as Bridge Engineer for the State Highway Commission of Kansas, and subsequently in the Bridge Department of the Missouri Pacific Railroad; also served for a short time as an assistant professor of applied mechanics at Kansas State College.

He was closely affiliated with the control of welding and related operations in the merchant shipbuilding program during the last war, serving upon committees of the American Bureau of Shipping and advisory committees of the War Metallurgy Committee and the U. S. Maritime Commission, for studying the behavior of various details of design and the effect of various welding procedures and the weldability of steel materials; also cooperating with shipyard personnel and naval architects on improvements in design practices and control of welding.

Harry E. Rockefeller

Harry Eugene Rockefeller was born in Asbury Park, N. J., in 1901. He graduated from Massachusetts Institute of Technology in 1922.

Upon his graduation, he was employed by The Linde Air Products Co. as a service operator in the sales department. Mr. Rockefeller has held various positions within the company, and is now Assistant Manager, Development Division. He has been intimately associated with most of the developments of The Linde Air Products Co., including such processes as flame-deseaming, flame-gouging, plate-edge preparation, and heliarc inert gas-shielded arc welding. Mr. Rockefeller also played an important part in the development of oxy-acetylene welding for pressure vessels, and the subsequent adoption of standards of design and construction for pressure vessels.

For many years, Mr. Rockefeller has been a member of the New York Section of the AMERICAN WELDING SOCIETY and has served on many of the Societies' committees. In addition, he has been active in the International Acetylene Association, Compressed Gas Association and American Society of Mechanical Engineers. He is the author of many papers on welding and related processes.

Mr. Rockefeller is with the New York office of The Linde Air Products Co., and makes his home in Bronxville, N. Y.

Harry E. Rockefeller

J. R. Stitt

J. R. (Ray) Stitt was born on Oct. 25, 1906, in Youngstown, Ohio. He was educated in the public schools in Lakewood and East Cleveland, Ohio, and graduated from Pennsylvania State College, with a B.S. in Civil Engineering in 1930.

J. R. Stitt

Upon graduation, he became Field Engineer for the Austin Co. In 1933 Mr. Stitt became Sales Engineer and Representative for Una Welding, Inc., with headquarters in Newark, N. J., and in 1935 was transferred to Unaweld Manufacturing Company of Texas with office in Houston. In 1937, he returned to the Austin Co. as Field Engineer and Inspector on the construction of welded buildings.

In 1938, Mr. Stitt was asked to go to Ohio State University to organize and teach the first complete curriculum in Welding Engineering. There he served as Assistant and Associate Professor for six years, during which time twenty-six students graduated with the degree of Bachelor Industrial Engineering (Welding Engineering).

From 1945 to date, Mr. Stitt has been Research and Welding Engineer of The R. C. Mahon Co., Detroit, Mich. The company fabricates and erects structural steel, rolling doors, steel roof deck insulated metal wall panels, industrial equipment and welded tanks, bases, etc.

Mr. Stitt is a member of the following engineering societies: AMERICAN WELDING SOCIETY, American Society for Metals, Chi Epsilon, The Engineering Society of Detroit, Society for Experimental Stress Analysis and Tau Beta Pi.

In connection with the AMERICAN WELDING SOCIETY, he has been a member since 1937 and has served on the Educational Committee since 1942. He was Chairman of the Detroit Section, 1948–1949 and Secretary-Treasurer of the Detroit Section since 1949.

DISTRICT REPRESENTATIVES 1950–51 NATIONAL NOMINATING COMMITTEE

District No. 1—Clarence E. Jackson

Clarence E. Jackson was born in Graceville, Minn. In 1927, he received a B.A. degree from Carleton College with distinction in physics. He continued graduate work at George Washington University at Washington, D. C. From 1930 to 1937, he was connected with the Metallurgy Division of the National Bureau of Standards in Washington. In 1937, he became associated with the Naval Gun Factory, transferring to the Naval Research Laboratory, in 1938, as Head of the Welding Section. The Secretary of the Navy awarded him the Distinguished Civilian Service Award with a citation for outstanding achievement in the field of welding. In 1946, he became associated with the Union Carbide and Carbon Research Laboratories, Inc. at Niagara Falls, N. Y., in his present position as research metallurgist doing research work in welding and related metallurgical fields.

Mr. Jackson assisted in preparation of the new *Welding Handbook* and has published over twenty articles dealing with welding and allied metallurgical problems. In addition to his membership in the AMERICAN WELDING SOCIETY, he is also

Clarence E. Jackson

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a member of the American Society for Metals, the American Institute for Mining and Metallurgical Engineers and an overseas member of the British Institute of Welding.

He is active on numerous Welding Research Council and local and national AMERICAN WELDING SOCIETY Committees. He is past-Chairman of both the Washington, D. C. and Niagara Frontier Sections of the AMERICAN WELDING SOCIETY.

District No. 3—D. H. Curry

D. H. Curry is a native of Tennessee. He attended Georgia School of Technology, Atlanta, Ga., obtaining a degree of B.S. in Civil Engineering. He later attended the Birmingham Center of the University of Alabama for studies of Metallurgy.

After graduation from Georgia Tech in March 1937, he was employed by Chicago Bridge and Iron Co. at their Birmingham Plant. He later held positions of timekeeper and engineer in the Erection Department for a short period, then returned to the Birmingham Plant, where he worked successively as Fitter, Layer-Out, Checker, Supervisor, Assistant Welding Engineer and was appointed Welding Engineer in 1947.

From 1942 to 1946, Mr. Curry served in the Civil Engineering Corps of the U. S. Navy as Supply Officer, First Lieutenant and Executive Officer aboard Floating Dry Docks in the Pacific.

He is a member of the AMERICAN WELDING SOCIETY and the American Society For Metals.

District No. 4—J. R. Stitt

(*For Biography—see Directors at Large above.*)

District No. 5—W. Gerard Fassnacht

Mr. Fassnacht was born at South Bend, Ind., Feb. 3, 1911. He was graduated from Purdue University with a B.S.Ch.E. in Metallurgy in 1932.

W. Gerard Fassnacht

Mr. Fassnacht's experience includes: 1933–1940, Bantam Bearings Div., Torrington Corp., Heat treater, Asst. Chief Metallurgist; 1940 to the present, Bendix Products Div., Bendix Aviation Corp., Metallurgist, Asst. Chief Metallurgist.

He is a charter member of the Michiana Section, A.W.S., having served as chairman and secretary-treasurer. He is also a member of the American Society for Metals, Tau Beta Pi, and Phi Lambda Upsilon.

District No. 6—A. K. Pandjiris

Anthony K. Pandjiris is President Pandjiris Weldment Co., St. Louis, Mo. He graduated from the Mechanical Engineering Dept. of Marquette University in 1938. Mr. Pandjiris served as Chairman, St. Louis Section, AMERICAN WELDING SOCIETY, 1949–50. He presented a paper at the annual meeting of the SOCIETY, Cleveland, Ohio in 1949. He was co-winner, James F. Lincoln Arc Welding Foundation Award in 1948.

District No. 7—Charles L. Haynes

For the past 25 years Charles L. Haynes has been with the C. F. Braun & Co. For several years, he was Chief Inspector. Later, he became Factory Superintendent. Mr. Haynes is now serving as a Staff Consultant assigned to the Construction Department.

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Metallurgy Class from Oklahoma A & M on visit to Beechcraft

Letter to the Society

Mr. J. G. Magrath, Exec. Sec.,
American Welding Society,
33 W. 39th St.,
New York 18, N. Y.

Re: Mr. W. H. Rice, Assoc. Prof.,
Oklahoma A & M College,
Stillwater, Okla.

Dear Joe:

I am sending you under this same cover a photograph of a group of students from Oklahoma A & M who recently made a two-day plant visitation trip to Wichita. They were accompanied by W. H. Rice, Associate Professor of Welding & Metallurgy and two of his staff. There were 18 or 20 boys in the group, all of which are not in the photograph.

This is at least the second year that I know of that Mr. Rice has brought his group to our city. Both years they have visited 6 to 8 of our different plants, both large and small. With their interest being primarily welding and metallurgy, the various plants visited have taken particular pains to show them all they could about their departments that would interest them most.

My point in bringing this to your attention is that each and every Section of A.W.S. should take it upon themselves to offer the colleges in their vicinity the opportunity to visit their local plants who are outstanding in welding. It would not be hard for the Section Chairman, Secretary or others to arrange for such plant visitation trips. A great many of these College Professors would jump at the opportunity if someone close to industry would eliminate the details of arrangement. It actually took me perhaps two hours to line up the "cream-of-the-crop" plants in our community. I am sure the public relations this accomplished for our city and local industry was well worth the effort and the good it did the boys might possibly be one of the turning points in their lives.

Sincerely,
R. L. Townsend
V. P. Dist. No. 6, A.W.S.

Annual Meeting

Elsewhere in this issue there is published the first tentative Program of the Annual Meeting of the American Welding Society to be held in Chicago during the week of Oct. 22, 1950. This Program constitutes the largest technical program in the history of the Society. Practically every phase of welding is covered by one or more of these papers. They represent the latest developments in apparatus, techniques, materials and procedure. These papers include also the results of hundreds of thousands of dollars of research work. Everyone intimately associated with the welding industry should make every effort to attend this Annual Meeting.

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TENTATIVE PROGRAM

AMERICAN WELDING SOCIETY

31st Annual Meeting
week of October 22, 1950

HOTEL SHERMAN, CHICAGO, ILL.

Welding and cutting demonstration will be featured at the National Metal Exposition held during the week in the Chicago International Amphitheatre

TECHNICAL SESSIONS

All Technical Sessions will positively start on time.

No Stenotype Reporter—Members desiring to discuss papers are urgently requested to prepare discussion in writing in advance of the meeting and to send copies to Headquarters, as those preparing written discussion will be given preference at the session. Members and guests giving extemporaneous discussion at meeting should forward a written discussion as soon as possible after the meeting.

SUNDAY AFTERNOON, OCTOBER 22nd

President's Reception

MONDAY MORNING, OCTOBER 23rd

Three Simultaneous Sessions

WELDABILITY

Some Observations on the Influence of Low-Temperature Cooling Rates Following Welding on the Ductility of Arc Welds in Mild Steel
by A. E. FLANIGAN, S. I. BOCARSKY and G. B. McGUIRE, University of California

The Effect of Nitrogen on Brittle Behavior of Mild Steel
by G. H. ENZIAN and G. J. SALVAGGIO, Jones & Laughlin Steel Corp.

The Micro-Mechanism of Tension Impact Fracture of Structural Steel—A Metallographic Study
by W. H. BRUCKNER, University of Illinois

Influence of Biaxiality on Notch Brittleness
by D. ROSENTHAL, University of California, and W. D. MITCHELL, Douglas Aircraft

ARCS AND ELECTRODES

The Nickel-Molybdenum-Vanadium Alloy Steel Shielded Arc Welding Electrode
by W. H. WOODING, Philadelphia Naval Shipyard

Aluminum-Magnesium Filler Metals for Welding High-Strength Aluminum Alloys
by R. D. WILLIAMS and D. C. MARTIN, Battelle Memorial Institute

The Rotary Electrode for Manual Metallic Arc Welding
by PROF. GILBERT S. SCHALLER, University of Washington

Research Techniques for Study of Arcs in Inert Atmospheres
by ROGER H. GILLETTE, The Linde Air Products Co., and R. T. BREYMEIER, Union Carbide and Carbon Research Labs., Inc.

SURFACING

Some Characteristics of Composite Tungsten Carbide Weld Deposits
by H. S. AVERY, American Brake Shoe Company

Hard Facing of Steam Valve Seats and Disks
by O. E. SWENSON, U. S. Naval Engineering Experiment Station

New Production Applications of Hard Facing
by ELDON C. HURT, Haynes Stellite Division, Union Carbide and Carbon Corp.

The Present Status of Composite Metal Fabrication by Arc Welding
by H. E. CABLE, The Lincoln Electric Company

MONDAY AFTERNOON, OCTOBER 23rd

Three Simultaneous Sessions

FLAME HARDENING

Commercial Flame Hardening
by E. J. COX, Pittsburgh Commercial Heat Treating Co.

Flame Hardening of Large Surfaces
by J. J. BARRY, Air Reduction Sales Co.

Production Flame Hardening
by MILTON GARVIN, Cincinnati Milling Machine Co.

ARCS, ELECTRODES & FLUXES

Effects of Reduced Atmospheric Pressures on the Characteristics of Arc Welding
by M. L. BEGEMAN, B. H. AMSTEAD and U. I. MASHRUWALA, Dept. of Mechanical Engineering, The University of Texas

Characteristics of Inert-Gas-Shielded Metal Arcs
by ALBERT MULLER, G. R. ROTHSCHILD, and W. J. GREENE, Air Reduction Co., Inc.

RESEARCH

Interrelation Between Mechanical Properties of Two Pressure Vessel Steels
by S. S. TOR, R. D. STOUT and BRUCE G. JOHNSTON, Lehigh University

Interpretation of the Work of the Fabrication Division, PVRC
by H. C. BOARDMAN, Chicago Bridge & Iron Co.

Sub-Critical Heating of Low-Carbon Plate Steel
by L. J. KLINGLER, Case Institute of Technology

PANEL DISCUSSION ON RECOMMENDED PRACTICES FOR RESISTANCE WELDING

SPONSORED BY A.W.S. RESISTANCE WELDING COMMITTEE

Bal Tabarin Room

Chairman—J. HEUSCHKEL
Co-Chairman—J. J. MacKINNEY

Historical Background
by WILSON SCOTT, Westinghouse Electric Corp.

Keynote of Quality
by R. T. GILLETTE, General Electric Co.

Aluminum and Magnesium Alloys
by W. J. WILSON, Lavelle Aircraft Corp.

Copper Alloys and Dissimilar Nonferrous Metals
by J. B. WELCH, Cutler-Hammer, Inc.

Carbon and Low-Alloy Steels
by J. HEUSCHKEL, Westinghouse Electric Corp.

Nickel and Nickel Alloys
by R. M. WILSON, Jr., The International Nickel Co.

Stainless Steels
by J. J. MacKINNEY, The Budd Co.

Coated Materials
by E. A. BUSSARD, Coleman Co., Inc.

High-Speed Motion Pictures

Summary and Adjournment

EDUCATIONAL LECTURE SERIES

MONDAY EVENING, OCTOBER 23rd

ADAMS LECTURE AND PRIZES

TUESDAY MORNING, OCTOBER 24th

Three Simultaneous Sessions

PIPE FABRICATION

Development of Welding Procedure for High Pressure, High Temperature Steam Piping
by N. L. NAVARRE, U. S. Naval Engineering Experiment Station

Fabrication and Welding of High Pressure-High Temperature Alloy Piping
by H. J. IRRGANG, Jr., W. K. Mitchell & Co.

Submerged Arc Welding to Pipe Fabricating
by S. R. JOSLIN, Welding Engineering, Inc.

Design and Installation of Large Diameter, Long Span, Welded Pipe
by HORACE JACKSON, Thompson Pipe and Steel Co.

RESISTANCE WELDING

A Practical Method for Obtaining Consistent Resistance Welds
by J. W. KEHOE, Westinghouse Electric Corp.

Variables in Cross-Wire Welding Dissimilar Metals
by I. S. GOODMAN, Westinghouse Electric Corp.

Resistance Spot and Seam Welding of Aircraft Materials Using DC Current
by J. H. COOPER, The Taylor-Winfield Corp.

The Projection Welding of 0.010- and 0.020-In. Steel Sheet
by E. F. NIPPES, J. M. GERKEN and J. G. MACIORA, Rensselaer Polytechnic Institute

SHIP STRUCTURE

Welded Reinforcement of Openings in Structural Steel Members
by D. VASARHELYI, and R. A. HECHTMAN, University of Washington

The Determination of Initial Stresses and Results of Tests on Steel Plates
by PROF. E. W. SUPPIGER, DR. CARLO RIPARBELLI and E. R. WARD, Princeton University

Some Metallurgical Aspects of Ship Steel Quality
by H. M. BANTA, R. H. FRAZIER and DR. C. H. LORIG, Battelle Memorial Institute

The Influence of Deoxidation Practice and Composition on the Toughness of Low-Carbon Steel
by J. F. WALLACE, Watertown Arsenal

TUESDAY AFTERNOON, OCTOBER 24th

Three Simultaneous Sessions

RESISTANCE WELDING

Optimum Flash Welding Conditions—The Importance of Upset Variables
by NIPPES, SAVAGE, PATRIARCA and McCARTHY, Rensselaer Polytechnic Institute

Flash Welding Nonferrous Materials
by F. L. BRANDT, JR., Thomson Electric Welder Co.

Spot Welding of Scaly Heavy-Gage Structural Steel
by NIPPES, RAMSEY and MACIORA, Rensselaer Polytechnic Institute

DESIGN

Special Luncheon

Introduction—H. R. MORRISON
Design for Production Economies
by G. F. NORDENHOLT, Editor, *Product Engineering*
Technical Session: Design for Production Economies
Organization for Design for Production Economy
by T. J. CRAWFORD, Consulting Engineer
Economies Accomplished by Redesign for Welding
by KENNETH JACKSON, Caterpillar Tractor Co.
Applications of Welded Design for Cost Reduction
by R. H. BENNEWITZ, The Linde Air Products Co.

SHIP STRUCTURE

The Underbead Crack Sensitivity of Some Shipbuilding Steels
by P. J. RIEPPEL and F. R. BAYSINGER, Battelle Memorial Institute

Stress Studies of Welded Ship Structure Specimens
by WM. R. CAMPBELL, Engineering Mechanics Section, National Bureau of Standards

The Distribution of the Locked-In Stresses in a Large Welded Steel Box Girder
by JOHN VASTA, Bureau of Ships, Navy Department

EDUCATIONAL LECTURE SERIES

TUESDAY EVENING, OCTOBER 24th—8:00 P.M.

Applied Welding Engineering

A Symposium on Proper Application of Welding Processes

SPONSORED BY A.W.S. TECHNICAL ACTIVITIES COMMITTEE

Bal Tabarin Room

Chairman—C. D. EVANS
Co-Chairman—S. A. GREENBERG

Introduction
by C. D. EVANS, International Harvester Co. (*Chairman*, Technical Activities Committee)

Design
by J. J. CHYLE, A. O. Smith Corp. (*Chairman*, Automotive Welding Committee)

Materials
by A. N. KUGLER, Air Reduction Sales Co. (*Chairman*, A.W.S. Committee on Brazing and Soldering)

Workmanship
by R. W. CLARK, General Electric Co. (*Chairman*, A.W.S. Committee on Standard Qualification Procedures and A.W.S. Army Ordnance Advisory Committee)

Inspection
by J. LYELL WILSON, Consulting Engineer (*Vice-Chairman*, Technical Activities Committee and *Chairman*, A.W.S. Committee on Welding in Marine Construction)

Applications
by S. A. GREENBERG, Technical Secretary, American Welding Society
Discussion

What are you welding today, Joe?
"PIPE LINES...and better than ever with G-E W-22 and W-52 ELECTRODES!"
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Like Joe and his friends, you, too, will find that G-E Types W-22 and W-52 electrodes have been recently improved for all positions, especially for down-vertical welding on pipe. Also recommended for structural frames, pressure vessels, and shipbuilding.
W-22 is an AWS E6010 electrode, reverse polarity d-c, all positions. It produces welds of excellent appearance, high tensile strength, ductility, and high impact resistance. W-52 is an AWS E7010 electrode, reverse polarity d-c, all positions. Recommended where a minimum of 70,000 psi is required; principally for welding low-alloy steels.
Regardless of what you are welding, the more than 70 types of G-E arc-welding electrodes assure you of just the right rod for your job. Your G-E Welding Distributor maintains a complete stock of G-E electrodes to meet your specific needs. Ask him for a copy of the handy pocket-sized G-E Electrode catalog, GEC-482. Or write: Apparatus Department, General Electric Co., Schenectady 5, N. Y.
ARC WELDERS • ELECTRODES • ACCESSORIES • RENEWAL PARTS
GENERAL ELECTRIC

WEDNESDAY MORNING, OCTOBER 25th

Three Simultaneous Sessions

RESISTANCE WELDING

Development of Production Welding Techniques
by J. RAYMOND WIRT, Delco-Remy Division, General Motors Corp.

Spot Welding of Thin Materials with Short Time-Constant Equipment
by FRANKLIN PAGE, JR., Du Fresne and Page Engineering Co., and FRANK JOHNSON, Unitek Corp.

Problems in the Design and Operation of Tube Mills for High-Speed Production
by DONALD H. FLEIG, American Electric Fusion Corp.

Contact Resistance
by DR. W. B. KOUWENHOVEN, The Johns Hopkins University

EDUCATION

SYMPOSIUM ON THE NEED OF PROFESSIONAL TRAINING FOR THE WELDING ENGINEER

Chairman—PROF. R. S. GREEN, Acting Chairman of the Department of Engineering, Ohio State University

Co-Chairman—PROF. M. L. BEGEMAN, University of Texas.

Employer Point of View:

Viewpoint of Representative of Bureau of Ships, Navy Dept.
by A. G. BISSELL, Bureau of Ships, Navy Department

Viewpoint of Representative of Industrial Research
by J. HEUSCHKEL, Westinghouse Electric Corporation

Viewpoint of Representative of Fabrication Industry
by J. J. CHYLE, A. O. Smith Corporation

Viewpoint of Representative of Private Research
by DR. J. M. PARKS, Welding Research, Armour Research Foundation.

College Point of View:

Viewpoint of Representative of a Western State University
by PROF. G. S. SCHALLER, Prof. Mechanical Engineering, University of Washington.

Viewpoint of Representative of an Eastern Private University
by DR. G. E. DOAN, Prof. of Metallurgy, Lehigh University.

Viewpoint of Representative of a Private Engineering School
by PROF. R. A. WYANT, Asst. Prof., Department of Metallurgy, Rensselaer Polytechnic Institute

Presentation of papers to be followed by an Authors' Round Table Discussion and General Discussion.

STRUCTURAL

Connections for Welded Rigid Portal Frames
by A. ANTHONY TOPRACTSOGLOU, LYNN S. BEEDLE, and BRUCE G. JOHNSTON, Lehigh University

Stress Redistribution Above the Elastic Limit for a Welded Frame
by L. E. GRINTER, Illinois Institute of Technology, and CHARLES G. PELLER, Valparaiso University

Buckling of Intermittently Supported Rectangular Plates
by C. H. NORRIS, D. A. POLYCHRONE and L. J. CAPOZZOLI, Massachusetts Institute of Technology

Welded Eccentric Girders for Composite Bridge Construction
by R. E. ROBERTSON, Consulting Engineer, Saxe, Williar and Robertson

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WEDNESDAY AFTERNOON, OCTOBER, 25th

BUSINESS MEETING

BOARD OF DIRECTORS MEETING

EDUCATIONAL LECTURE SERIES

WEDNESDAY EVENING, OCTOBER 25th

SECTION OFFICERS CONFERENCE

UNIVERSITY RESEARCH CONFERENCE

THURSDAY MORNING, OCTOBER 26th

Three Simultaneous Sessions

NONFERROUS METALS

Welding Aluminum with Inert Arc D. C.
by JOHN W. MORTIMER, Whitlock Mfg. Co.

The Effect of Welding on the Properties of Titanium-Carbon Alloys
by E. M. MAHLA and R. B. HITCHCOCK, E. I. duPont de Nemours & Co.

Aircomatic Welding of Copper-Base Alloys
by HAROLD ROBINSON and J. H. BERRYMAN, Air Reduction Co., Inc.

Copper Alloy Brazing for Production Economy
by CLINTON E. SWIFT and E. B. BROWN, The American Brass Co.

PRESSURE VESSELS

Stress Concentration Problems in Welded Construction at Atmospheric and Elevated Temperatures
by G. J. GREEN and D. H. MARLIN, Dravo Corp.

Multiple Layer Submerged Arc Welding of Pressure Vessels
by L. C. STILES and D. H. CURRY, Chicago Bridge & Iron Co.

General Procedure for Fabrication of Pressure Vessels
by S. V. WILLIAMS, A. O. Smith Corp.

Welding of Stainless Steel Containers To Meet Sanitary Standards
by P. H. MOUNTS, Cherry-Burrell Corp.

AIRCRAFT

Heliarc Welding in Production
by T. E. PIPER, Northrup Aircraft, Inc.

The Evaluation of Factors Influencing the Crack Sensitivity of Several Aircraft Steels
by A. W. STEINBERGER, B. J. DeSIMONE and J. STOOP, Curtiss-Wright Corp.

Some Basic Problems in Design for Aircraft Welding
by J. KOZIARSKI and K. B. GILLMORE, Piasecki Helicopter Corp.

Jigs and Fixtures for Inert Gas Arc Welding
by H. A. HUFF, JR., and A. N. KUGLER, Air Reduction Sales Co.

THURSDAY AFTERNOON, OCTOBER 26th

Inspection Trip

FRIDAY MORNING, OCTOBER 27th

Three Simultaneous Sessions

INERT-GAS METAL-ARC WELDING

Aircomatic Welding of Nickel and Stainless Clad Alloys
by L. W. WILLIAMS, Lukens Steel Co.

Aircomatic Welding of Austenitic Nickel-Chromium Stainless Steel
by W. G. BENZ, JR., and J. S. SOHN, Air Reduction Co., Inc.

Argon Metal Arc Welding Nonferrous Metals and High- and Low-Alloy Steels
by H. T. HERBST, The Linde Air Products Co.

Construction of Welded Aluminum Ammonium Nitrate Prilling Tower
by H. N. HOCKENSMITH, Brown & Root, Inc.

OXYGEN CUTTING

The Use of Electronically Guided Template Tracing Devices in Shape Flame Cutting
by R. O. FISH, Fairbanks, Morse & Co.

Foundry Applications for Oxygen, Acetylene and Other Gases
by G. E. BELLEW, Air Reduction Sales Co.

Production Edge-Preparation and Shaping of Plate for Welding
by C. A. HEFFERNON, The Linde Air Products Co.

Design Around a Process
by R. F. HELMKAMP, Air Reduction Sales Co.

LOCOMOTIVES AND RAILROADS

Fabrication of Main Line and Switcher Locomotives
by FRED T. PERRY, General Electric Co.

Reclamation of Diesel Locomotive Parts
by LaMOTTE GROVER and R. L. REX, Air Reduction Co., Inc.

Economy of Reclamation by Welding at the Railroad Reclamation Plant
by W. G. MUSCHLER, Chicago, Burlington & Quincy Railroad

Rehabilitation of Ore Cars
by J. F. LIKARISH, Great Northern Railway Co.

FRIDAY AFTERNOON, OCTOBER 27th

Three Simultaneous Sessions

MACHINERY AND MAINTENANCE

Welding as Applied to Manufacture of Farm Tractors
by J. L. BUCHHOLZ, International Harvester Co., Farmall Works.

Tubular Sections in Frame Design of Hydraulic Benders
by E. J. DeWITT, Wallace Supplies Mfg. Co.

Electric Utility Welding in the Shop and Field
by DAVID P. O'CONNOR, Department of Water and Power, City of Los Angeles, Calif.

CONTROLS, STANDARDS

Procedure Approval Tests
by S. S. KATSEF, U. S. Naval Engineering Experiment Station

Better Welding at Lower Costs
by LEW GILBERT, Editor, *Industry and Welding*

Suggested New Welding Standards
by J. F. LINCOLN, The Lincoln Electric Company

MACHINE WELDING

Welding with Multiple Electrodes in Series—a New Unionmelt Method
by E. L. FROST, Union Carbide and Carbon Research Labs., Inc.

Welding Characteristics of Submerged Arc with Three-Phase Power
by N. G. SCHREINER, The Linde Air Products Co., and E. A. CLAPP, Union Carbide and Carbon Research Labs., Inc.

Automatic Welding in Steel Mill Maintenance
by W. P. HOFFMAN, International Harvester Company, Wisconsin Steel Works

NEWS OF THE INDUSTRY

A.S.T.M. Symposium on Effects of Low Temperatures on the Properties of Materials

This Symposium on Effects of Low Temperatures on the Properties of Materials, originally presented at a meeting of the Philadelphia District of the American Society for Testing Materials in March 1946, and reviewed by the authors in October 1949, has just become available in printed form (A.S.T.M. Special Technical Publication No. 78) and covers specifically the fields of both ferrous and nonferrous metals, plastics, and rubbers. Much interest is focused on this subject because of problems in refrigeration, the chemical processes, etc. The purpose of this Symposium is to bring together some of the existing facts, data, and experience, and to suggest what might be worked for in the future.

Following an Introduction by A. O. Schaefer, the publication includes four profusely illustrated papers: "Low Temperature Behavior of Organic Plastics," H. K. Nason, T. S. Carswell, and C. H. Adams; "Low Temperature Properties of Elastomers," John W. Liska; "Effects of Subatmospheric Temperatures on the Properties of Nonferrous Metals," Norman L. Mochel; "The Properties of Weldments at Low Temperatures," Robert D. Stout.

Copies of this 68-page booklet, heavy paper cover, can be obtained from the American Society for Testing Materials, 1916 Race St., Philadelphia 3, Pa., at $1.50 each.

Rensselaer Polytechnic Institute Expands Instruction in Design for Welding

The Structural Engineering Division of the Civil Engineering Department of Rensselaer Polytechnic Institute has recently expanded the sequence of courses, which must be taken by all students electing the structural option, with the addition of a new course to permit a better coverage of design for welding. A structural option student is now required to take ten courses in structural analysis and design totaling eight hundred contact hours.

Welded beam connections, including moment resistant connections, and miscellaneous welded joints, as well as a welded roof truss are included in a design course taken by all Civil Engineering students in the second term of the Junior year. The analysis of groove welds by the principal stress method is included. This course has been included in the curriculum for several years.

The new course, taken by Structural option men in the second term of the Senior year, includes the design of a welded rigid frame for a building. This follows a first term Senior course of eighty contact hours in the analysis of Indeterminate Structures and is taken coincidentally with another course of eighty contact hours in Indeterminate Structures. It is therefore unnecessary to teach methods of analysis in connection with the design. The frame is first analyzed by the general method and is then checked by moment distribution. It is analyzed with column bases pinned and then fixed to determine effects of fixation at column bases. Temperature changes, girder shortening due to axial loads and the effects of the inclusion or omission of a tie are considered. It is also analyzed with and without haunches to determine haunch effect. Bracing is designed as well as all welds and a drawing is made of the complete design.

Mr. Martin Korn, Consulting Engineer of Dearborn, Mich., has given valuable assistance in the form of blueprints of existing rigid frames and advice based on a wide experience in the design of this type of structure. Mr. Korn has expressed a willingness to give similar assistance to other schools interested in improving existing courses in design for welding with the introduction of this type of design problem providing these schools show sufficient initiative to launch such courses.

I. A. A. Convention

The 50th Annual Convention of the International Acetylene Association, held at the Fairmont Hotel in San Francisco, on March 27–29, was an outstanding success. This represented the first convention held on the Pacific Coast.

The opening lunch on March 27th was attended by 300 members and guests. Following a warm welcome to San Francisco by the Honorable Elmer E. Robinson, its genial mayor, President A. J. Fausek gave an inspiring talk on "What Constitutes our Security." The Award of the James Turner Morehead Medal for 1949 to James I. Banash was one of the highlights of the luncheon. The Honorable Karl R. Bendetsen, Assistant Secretary, Department of the Army, was guest speaker. Mr. Bendetsen spoke on "International Realities."

A number of members were elected as Honorary Members of the I.A.A. Members of the AMERICAN WELDING SOCIETY will recognize in most of these individuals prominent members of the SOCIETY. These Honorary Members are: Worthy Channing Bucknam, Glenn O. Carter, David Schenk Jacobus, Philip Kearny, Albert McMillan, Alvah R. Small, H. Sydney Smith, Herman Van Fleet and Robert Allen Witherspoon.

The presentation by T. C. Fetherston and the premier showing of a new motion picture, "The Oxyacetylene Flame—Master of Metals," was given at the opening session on Monday afternoon, March 27th. This basic sound, color, motion picture of the oxyacetylene flame has been excellently prepared by Transfilm, Inc., and is going to fill a long felt need. The reaction of those in attendance was very gratifying and only favorable comments were heard.

The technical sessions were well attended.

Officers Elected

The following Officers were elected by the National Welding Supply Assn. to serve during the coming year:

President—Glenn G. Garman, Purity Cylinder Gases, Inc., Grand Rapids, Mich.

Vice-Presidents— R. S. McCracken, Jr., R. S. McCracken & Sons, Inc., Philadelphia, Pa.; William A. Rice, Virginia Welding Supply Co., Charleston, W. Va.; Charles O. Stilwell, Gulf Welding Equipment Co., New Orleans, La.; Wm. F. Johnson, Johnson Welding Equipment Co., Inc., Chicago, Ill.; Walter G. Lofgren, United States Welding Works, Inc., Denver, Colo. and G. Milton Sims, Absco Welding Supplies, Los Angeles, Calif.

Board of Directors (Term Expiring 1953)—Homer W. Cessna, Cessna Welding Supply Co., Baton Rouge, La.; R. P. Tarbell, Scott-Tarbell, Inc., Cleveland, Ohio; J. Gordon Duff, Gordon Duff, Los Angeles, Calif. and R. P. Mahony, Albany Welding Supply Co., Albany, N. Y. (Term Expiring 1952 to fill the unexpired term of R. S. McCracken, Jr. who was elected Vice-President), Elmer Bakke, Welders Supply Co., Inc., Seattle, Wash.

Bassler Moves Office

The Taylor-Winfield Corp. of Warren, Ohio, announces a change of address of their Chicago representative, The Denton & Anderson Co. The Chicago district office under Clyde G. Bassler has been moved from 543 W. Washington Blvd. to Rooms 510 and 512 at 9 S. Clinton St., Chicago 6, Ill. He is a member of the AMERICAN WELDING SOCIETY.

Resistance Welding Machine Shipments Up 23%

Shipments of resistance welding machines took a definite upturn in March according to the Resistance Welder Mfg. Assn. As compared to February, they increased 23%. At the same time the Industry's backlog of unfilled orders on hand on March 31st advanced 16% over the end of February.

This Lukens man *knows* . . .

Use "140" MONEL electrodes for welding the clad side of Lukens Monel-Clad Steel

"When customers ask us to recommend a welding rod for the clad side of our Monel-Clad Steel, we tell them to use the new '140' Monel electrode."

These are the words of Mr. L. W. Williams, Manager, Technical Service, Lukens Steel Company, Coatesville, Pa. Mr. Williams' expert advice on metal problems is sought by many users of his company's products.

In explaining his reasons for recommending "140" Monel® electrodes, Mr. Williams said:

"The '140' Monel electrode was especially designed to lay its deposit directly on steel. Since the ductility of the deposit, arcing characteristics and slag removal are unaffected by steel pick-up, there is no need for the customary barrier layer of Nickel.

"Moreover, we know that the customer can be sure the welds he makes with the '140' Monel electrode will meet various code requirements and pass X-ray inspection."

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In addition to welding the clad side of Lukens Monel-Clad Steel, "140" Monel electrodes give good results in any Monel welding application where steel is present . . . making Monel overlays on steel, joining Monel to steel or stainless steel.

"140" Monel electrodes are available in standard diameters of 3/32", 5/32", and 3/16". Current requirements are – D.C., reversed polarity.

For additional information about the welding of Inco Nickel Alloys, write for: *Bulletin T-2, "The Welding, Brazing, and Soldering of Monel, Nickel, and Inconel."*

NOW—FOR YOUR CONVENIENCE—Inco welding electrodes are now supplied in new containers . . . easier to open, better protection during shipping and storage.

THE INTERNATIONAL NICKEL COMPANY, INC.
67 Wall Street, New York 5, N. Y.

CUT WELDING COSTS 263%* with

Norelco CONTACT WELDING ELECTRODES

• With this entirely *new type* of rod, welding speeds are increased an average of 45%. Welding is made much easier by touch-welding in all positions; and quality uniformly improved through less distortion, undercutting, and better control.

Available in two types: Contact 18 (in conformance with AWS E6013 class) for fast all-position general-purpose welding, and Contact 20 (AWS 6020 class) for extremely fast production welding in horizontal and downhand positions. Both AC or DC.

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*For full details of actual production results, technical information and prices write:

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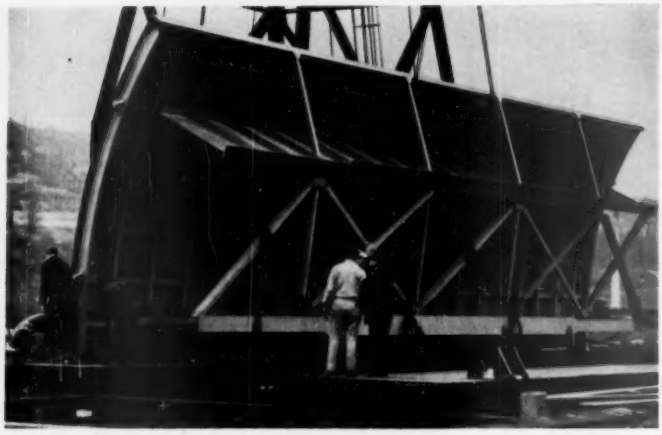

Welded Tainter Gate

Tainter Gate

One of six big tainter gates to be installed on the new Monongahela River dam at Morgantown, W. Va., is shown here at the Dravo Corp. shipyard, Pittsburgh, being prepared for loading aboard a barge. Fabricated of welded steel plate, the 65-ft. long gates are to control flow of water through the concrete dam spillways. The dam and accompanying lock are being built by The Contracting Division of Dravo Corp. for the U. S. Army Corps of Engineers as part of a plan to improve navigation facilities on the Monongahela for increased transportation of coal downstream to Pittsburgh's industrial area. Besides the six tainter gates, Dravo's Engineering Works Division also is fabricating six 70-ft. long plate girder service bridges to span the piers of the dam.

Wasserman Luncheon

Eutectic Welding Alloys Corp. arranged a luncheon, Friday May 19th, at the New York Athletic Club in honor of Hughes Wasserman, welding pioneer and father of R. D. Wasserman, president of the Company, on a visit here from Switzerland.

The elder Wasserman has been a metallurgist and pioneer in the welding field all his life, having started his career as assistant to Dr. Goldschmidt, the world-renowned inventor of the Thermit welding process. In 1906, he opened a laboratory for welding research and manufacturing of welding compounds, in Lausanne, Switzerland. For the past 44 years he has been President of Societe des Soudures Castolin, S. A., Lausanne, which produces Eutectic Products in Europe.

Mr. Wasserman was one of the founders of the Swiss Welding Society and was well acquainted with the erstwhile pioneers of oxyacetylene welding.

Today, at 76, he is still active, and made his first trip to the United States.

Phillip E. Graham Co., Inc., Moves to Larger Quarters

Phillip E. Graham, Inc., of Indianapolis, Ind., sales representative for Acro Welder Mfg. Co., has recently moved to larger quarters, approximately 7000 sq. ft. in order to take care of the constantly expanding business.

The Graham Co. carries a complete stock of perishable parts for all makes of Resistance Welder Controls and rebuilds these Controls at their Plant, or gives service at the customer's plant.

They also carry a complete line of all makes of Industrial Electronic Tubes. Mr. Graham is a member of the AMERICAN WELDING SOCIETY.

Robt. W. Hoffman and P. Arthur Green Merge

The Robt. W. Hoffman Co. of Chicago, Ill., distributors of Acme & Acro Spot Welders, has announced a recent reorganization and expansion program in the field of Resistance Welding.

Robt. H. Hoffman and P. Arthur Green, formerly with Chicago Welder & Machine Co., Hinsdale, Ill. have consolidated their companies to offer a combined welding sales and service organization. The new showroom, welding laboratory and shop is located at 32-34 S. Clinton St., Chicago.

Messrs. Hoffman and Green have made several contributions in the form of resistance welding tools in the past few years. One outstanding contribution being the Butt-On-Spot butt welding attachment for spot welders, and more recently, the Weld-Air-Matic packaged air conversion unit for spot welders.

The firm will operate under the name of Robt. W. Hoffman Co., Inc. with Robt. H. Hoffman as President and Treasurer, Corinne V. Hoffman as Vice-President and P. Arthur Green as Secretary. Both are members of the AMERICAN WELDING SOCIETY.

NEW LITERATURE

Welded Shovels

The April issue of the "Marion Groundhog" house organ of Marion Power Shovel Co., Marion, Ohio, contains a series of articles on welding in connection with Marion products.

Nickel Technical Bulletin

A new technical bulletin has recently been issued by International Nickel on the resistance of nickel and its alloys to corrosion by caustic alkalies. Twenty-four pages in size, it contains over 20 illustrations and over 40 tables. Methods of testing as well as results are given. Nickel-alloy irons as well as the higher nonferrous nickel alloys and pure nickel are discussed.

Performances of the materials are quoted in a variety of fields including soda manufacture, transportation and production of such products as viscose rayon, soap, pulp and paper. The field of petroleum refining is covered, and instructions are included for descaling and cleaning for maintenance welding.

Designated as Technical Bulletin T-6, the publication is available without charge through The International Nickel Co., Inc., 67 Wall St., New York 5, N. Y.

Arc Welding at Work

Volume VII, No. 1 "Hobart Arc Welding News," a 24-page booklet of interesting photographs and articles on welding from all over the country, is now available. Copies are mailed free of charge to anyone interested in arc welding.

Many of the articles feature time and money saving applications. To get your copy, write the Hobart Brothers Co., Troy, Ohio.

Hard-Facing Electrodes

Stoody Co., manufacturers of hard-facing alloys, Whittier, Calif., recently announced new a.-c. d.-c. coatings for several of their electrodes formerly limited to d.-c. application only. Among these are Stoodite, Tube Borium and Borod. The a.-c. d.-c. coated rods do not replace the older d.-c. types, but are additions to the regular Stoody line. It becomes necessary, therefore, to distinguish which rod is desired when ordering. To differentiate between the two types, the new rods are preceded by the term "a.-c.-d. c." The original d.-c. electrodes will continue to be specified by their established names, as in the past.

Literature fully describing the characteristics of the new a.-c. d.-c. Stoody electrodes is available from the manufacturer.

Ampco Brochure

In a new colorfull 4-page brochure Ampco Metal, Inc., Milwaukee 15, Wis., offers information regarding the most active items of their Ampco Weld line of Resistance Welding tips, wheels, water-cooled holders, welding dies, welder shafts, bushings and quick disconnect water shut-off couplings. Copies of this Bulletin RW-2 will be supplied upon request.

Flash Welding—Without Flash

Explanation of how percussion welding techniques have been successfully adapted to miter welding extruded aluminum sections without producing any flash is contained in the lead article in "Technical Advisor No. 13" just issued by Reynolds Metals Co., 2500 S. Third St., Louisville 1, Ky.

The article details results of a study made to reduce flash resulting from miter welding extruded aluminum sections such as those used in the manufacture of window frames, screens, storm windows and the like.

A newly devised flash welding system is described which permits the flashing period to be reduced appreciably while also confining the pushup to a hammer blow to obtain the advantages of percussive welding.

Also dies gripping the work are relieved by cutting them back at about 60°, permitting a closer grip on the work and exertion of a "pinching" action to inch off the flash.

As a result, upset metal that originally extended for $\frac{1}{4}$ in. across the weld zone to create a tremendous amount of work in finishing off the weld has been reduced to a mere line, practically eliminating any need for extensive weld finishing. The weld zone is reduced to less than one-hundredth of an inch, with no weakening of metal adjacent to the weld zone from annealing.

This latter feature is important to aluminum fabricators because it means that high-strength aluminum alloys can be fabricated in the heat-treated condition with hardly any reduction in strength due to welding. Previously, where strength was important, it was necessary to heat treat the fabricated assemblies after welding, an expensive and sometimes impractical operation with large or bulky weldments. The new technique eliminates this difficulty.

For complete details on the new process, write for "Technical Advisor No. 13," sent without charge. Send your request to Reynolds Metals Co., 2500 S. Third St., Louisville 1, Ky.

Welded Open-Hearth Ladles

The William B. Pollock Co., Youngstown, Ohio, has issued a four-page brochure on their welded open-hearth ladles.

High-Temperature Steam Piping

The American Bureau of Shipping, 45 Broad St., New York 4, N. Y., has issued in mimeographed form "An Interim Guide for the Installation of High-Temperature Steam Piping on Ships."

This report was recently prepared by the Bureau with the cooperation of an Advisory Panel drawn from the oil refineries, central power plants and the marine industry.

Regulator Catalog

National Welding Equipment Co., of 218 Fremont St., San Francisco, Calif., have released an attractive catalog fully descriptive of a well-illustrated line of precision regulators for the medical field and for use with the rare gases as employed, for instance, in the atomic hydrogen and helium or argon arc-welding processes.

Among the features particularly stressed by the manufacturer are a flowmeter with a suitably treated background permitting night reading, self-reseating relief valves for both the primary and secondary stage-reduction chambers for maximum safety and a new development which eliminates friction of the pressure adjusting mechanism.

Temperature Measurement

Tempil° Corp. announces that a new brochure describing its line of temperature-indicating products is now available.

Directions for use and typical applications for Tempilstiks°, Tempilaq° and Tempil° Pellets are listed. Available temperature ratings cover the range from 113 to 2500° F. in closely spaced temperature intervals.

Copies of the new brochure can be had by asking for Catalog No. 501.

Contact Type Electrode

A recently published folder on the rapidly increasing use of the new super-fast electrode Hand-Omatic for steel, is available free from Eutectic Welding Alloys Corp., 40 Worth St., New York 13, N. Y.

A series of simple experiments are shown in the folder, illustrating the ease with which Hand-Omatic can be used. For instance, any welder can run a perfect bead,

REGO

TWO-STAGE
REGOLATOR

Precision Regulation
for
OXYGEN
ACETYLENE
HYDROGEN

The precision performance of the two-stage RegOlator is especially desirable for piped distribution systems and other applications where large volumes of high pressure gases must be controlled accurately.

Two-Stage RegOlator Gives You Plus Performance!

Constant delivery pressure regardless of drop in pressure at the inlet...Patented design incorporates nozzle-type first stage counterbalanced by stem-type second stage... Triple action cartridge filter eliminates troubles usually caused by rust, dust or dirt.

Listed by Underwriters' Laboratories, Inc. and Factory Mutual.

Write for complete information.

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The BASTIAN-BLESSING Company
4201 West Peterson Ave. Chicago 30, Illinois

PIONEER AND LEADER IN THE DESIGN AND MANUFACTURE OF PRECISION EQUIPMENT FOR USING AND CONTROLLING HIGH PRESSURE GASES

sight unseen, on the under side of a steel plate. Or, the electrode holder can be placed on a steel sheet with the electrode at a 20° angle; the latter will actually burn itself down without any guidance by the hands!

Propane Cylinders

Harrisburg Steel Corp. of Harrisburg, Pa., pioneer manufacturer of gas cylinders in this country, has issued a new four-page three-color folder on its "Lite-Weight" Propane Cylinders.

Illustrated are I.C.C. Specification 4BA-240 cylinders of 100 lb. capacity and 72 lb. tare weight, now furnished in a choice of aluminum or red oxide ground coats, with or without caps and valves inserted, and with customer's registered mark and serial numbers at no extra charge.

The folder also illustrates four principal manufacturing operations on the cylinders: automatic welding with submerged arc, cold drawing heads in mechanical press, annealing furnace operation and valving and loading cylinders into cars via a new bracing and blocking technique which gets 500 to 550 cylinders into a single ordinary freight car.

Persons desiring this folder are asked to request Bulletin 466 from Harrisburg Steel Corp., Harrisburg 31, Pa.

Ampco News

Learn how one concern successfully and economically welded a galvanized recirculation air duct system; how cast-iron tumbler gears were salvaged or how steel dies overlaid increased die life 400% while materially improving press room efficiency.

Possibly your screw conveyor shafts have gone to pot and need rebuilding; or you are interested in fabricating an Ampco Metal fractionating tower; or could get over a tough spot with some specially designed spot welding arms; all this information is contained, briefly stated, in the spring issue of Ampco Welding News. A copy of this informative house organ will be sent gratis. Just write Ampco Metal, Inc., Milwaukee 15, Wis.

Chart Available on Tests for Identifying Metals

A chart entitled "Simple Tests for Identifying Metals—To Aid in the Determination of Correct Oxyacetylene Welding Procedure," is available from the International Acetylene Association. The method of identifying 13 different metals and alloys—including the cast irons, low-carbon steel, high-carbon steel, copper and its alloys, aluminum and its alloys, nickel and lead—by appearance, spark test, chip test and blowpipe test are explained in tabular form. The chart measures 15 by 11 in.

Copies of this chart will be sent to those requesting it, as long as the supply lasts. Requests should be addressed to the Secretary, International Acetylene Assn., 30 E. 42nd St., New York 17, N. Y.

NEW PRODUCTS

Arc-Welding Electrode

A new improved arc-welding electrode, the W-52 (A.W.S. Class E7010) a reverse polarity d.-c. rod, has been announced by the Welding Divisions of General Electric's Apparatus Dept.

The new W-52 electrode is available in $^{5}/_{32}$ and $^{3}/_{16}$ in. diameters for field trial. This carbon-molybdenum electrode is specifically designed for high-quality welding of low-alloy, high-tensile steels, such as pipe lines, in all positions. It can be widely applied in the welding of high-pressure piping and of castings where high-tensile strength and resistance to creep at high pressures and temperatures are desired.

Good bonding action at the fusion zone on horizontal fillets and less tendency to produce pin holing on vertical down welds are characteristics designed into the new electrode.

The stable arc of the new W-52 with low spatter lends for smooth operation and eliminates tendencies to short out in confined joint preparations.

Cleaning Solution

A new cleaning solution for personal safety equipment such as goggles, respirators and helmets is announced by American Optical Co., Southbridge, Mass.

Combining the best properties of both a germicide and detergent, the No. 101 Germicidal Detergent removes dirt and deposited skin oils without inducing corrosion, staining or deterioration of rubber, plastic or metal parts. The solution may be used as a spray, swabbing or immersion solution.

Odorless, the No. 101 also acts as a deodorant in certain cases. Independent laboratory tests show that the solution is better than the general types of "sanitizing" fluids such as coal tars, cresol compounds or formaldehyde, the company reports. Tests also proved the new product, the company said, to be neither a primary nor a secondary irritant, even after repeated usage.

The fluid is effective in hot or cold water, although increased temperatures will tend to aid its effectiveness and reduce cleansing time.

Available in gallon containers or 2 oz. bottles packed 10 to a carton, a gallon of the solution makes 64 gallons of cleanser of a greater strength than generally recommended for disinfecting.

Silver Joining Alloy

A new silver joining alloy with a remelt temperature of 1850° F., plus electrical conductivity equal to twice that of conventional silver solders, is announced by the Eutectic Welding Alloys Corp., 40 Worth St., New York 13, N. Y.

Demand for this new alloy—called EutecRod 1807—was created by the need for an alloy which would withstand normalizing and heat-treatment temperatures. Only a remelt temperature in the range of 1800 to 1900° F. would satisfy these conditions.

Corrosion resistance of EutecRod 1807 is excellent, hardness is 64–70 Brinell, tensile strength is 30,000 to 32,000 psi. Sizes available are $^{1}/_{64}$", $^{1}/_{32}$", $^{1}/_{16}$" and $^{1}/_{32}$-in. coils. A bulletin with more information is available from the above address.

A.-C. Transformer Welders

Hobart Brothers Co., Troy, Ohio, announces a complete new line of transformer type a.-c. arc welders which have no moving parts. By means of an electrically adjustable reactor, the usual laborious cranking method of setting welding current has been eliminated. Two models are now in production and available for delivery. Model TSP-205-C with power factor correction is rated 200 amp. at 30 v. Model TSP-182-C with power-factor correction and limited input for operation on REA lines is rated 180 amp. at 25 v.

Control of the five main taps of welding current from low to maximum is accomplished with a large, easy to move hand wheel switch. The switch is heavy copper, molded in bakelite and is mounted on the front panel. Above this hand wheel is the rheostat control knob, which controls the amount of d.-c. current flowing into the reactor to give 100 fine welding current settings in each of the five main steps or ranges. Thus, approximately 500 different welding current settings are available without changing electrode or ground cable connections, or without laborious cranking.

In addition to adjusting the welding current by changing the relationship between the d.-c. flux and the a.-c. flux in the reactor core, the rectified (d. c.) current has the effect of regulating this relationship at the time the arc is struck to provide a higher than normal current at the arc for a fraction of a second. This gives the operator an instantaneous striking arc. Easy arc striking is therefore inherent in the Hobart design, without involving any moving parts.

By connecting the rheostat to the reactor circuit through a standard plug and receptacle, remote control is made possible on all models of these new welders. The rheostat can easily be removed and used at a distance from the welder by merely inserting an ordinary extension cord with standard plug and receptacle.

A toggle-type line switch is provided on the control panel for disconnecting the transformer from the power line. Also, because of the extreme quiet operation of these welders, a single light is provided on the panel to signify that the set is on or off.

Model TSP-205-C is designed to operate on single phase 60 cycles either 220/440 dual voltage or 550 volts. Can also be supplied for 50 cycles. Model TSP-182-C is designed to operate on single phase, 60 cycles, 230 v., to comply with the REA specifications for welders used on their 3-kva. transformers. Steel casters are standard equipment.

Electrode for High-Speed Production

A new, easy operating, mild steel E6012 electrode, "Fleetweld 72," is introduced by The Lincoln Electric Co. of Cleveland, Ohio. For high-speed production of single-pass fillet and lap welds in the flat or horizontal positions, "Fleetweld 72" produces outstanding welding economies.

The new electrode operates best, without overheating, at currents above those normally used with E6012 electrodes. With higher melt-off rates, greater welding speeds are obtained in depositing dense, high strength, ductile welds. Arc characteristics secure excellent root penetration. The near flat (slightly convex) bead eliminates piling up of excess weld metal usually encountered with E6012 electrodes, thus gaining greater weld length deposited per electrode.

Operating characteristics of the electrode make it an exceptionally easy one to use. The operating current range is broader than that of conventional E6012 electrodes and the type of arc is especially suited for working where fit-up is poor. It also is ideally suited to making lap welds in which the weld feathers smoothly into the top edge of the joint. Unusually high-production speeds have been obtained with this electrode on lap weld applications. The smooth flowing arc deposits

welds with a minimum of spatter, even at high amperages. Slag is practically self-cleaning.

This new premium performance electrode is available at standard prices in 14 and 18 in. lengths and in the following diameter sizes: 1/8, 5/32, 3/16, 7/32, 1/4 and 5/16 in.; for operation on both a. c. and d. c.

Insulated Electrode Holder

An innovation in the insulated electrode holder field is introduced by The Lincoln Electric Co. of Cleveland, Ohio. The new holder, the Lincoln LJ-1, designed around an unusually small size, combines the advantages of both tong and hole type holders in providing a strong, "lock-jaw" electrode grip yet easy thumb-pressure electrode release. The holder's small size is ideal for operation with heavy gloves and in corners and other confined areas. The small size, lightweight and perfect balance make an exceptionally comfortable holder to operate.

New materials have been developed especially for the holder to provide strength and ruggedness as well as insulating qualities. The tip of the jaw end of the holder is covered with a new, high-temperature resistant, asbestos base compound which is highly flexible. The tip cannot be broken by the impact or shock of dropping or crushing. The tip will also resist the abrasion of rubbing against metal as well as the wear caused by weld spatter. Weld spatter will not adhere to it.

Because of its flexibility, the molded tip can be formed to cover the end of the holder completely, thus giving full insulation. The electrode is held close to the end of this tip so that the holder can be worked in close quarters without danger of grounding. The handle and jaw lever arm are also completely insulated, covered with a tough high temperature fiber.

The metal frame of the holder is a new material especially developed to give the best combination of strength and thermal conductivity. It is a high-strength electrical bronze.

A lever type jaw is used for gripping the electrode. The jaw is loaded with 50% additional spring capacity which gives approximately nine times as much grip as coil springs. The release handle lever, however, gives a high degree of leverage so that inserting or releasing an electrode is quick and easy. A light thumb pressure on the arm opens the jaw. The cam jaw construction, open to one side of the holder, gives one position, 20° off vertical or downhand welding, and a second position, 30° off horizontal for overhead welding. A hole is provided in the jaw for holding 18-in. center grip electrodes.

The Lincoln LJ-1 is manufactured in the 250-amp. size for 1/16- to 5/16-in. electrodes.

Cover Lens

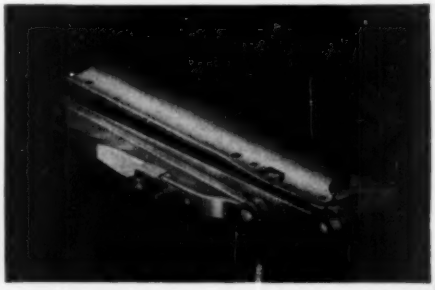
The Thermacote Mfg. Co. of New Jersey and Los Angeles, manufacturer of coated-welding cover lenses, announces the availability of their new All Plastic Welders Cover Lens.

The special plastic used in the new Thermacote All Plastic Cover Lens is extremely clear and will not warp or crack.

A.-C. Welder

A new silicone-insulated, portable, a-c welder known as the 6WK20H series has been announced by the Welding Divisions of General Electric's Apparatus Department. The 6WK20H series are part of the only standard a-c welder line commercially available incorporating silicone insulation. A high margin of safety and operating dependability is provided by this insulation since it is unaffected by high temperatures and is water repellent.

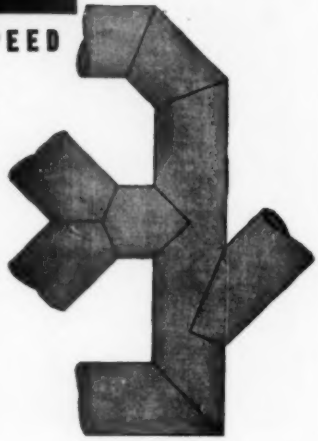
Compact construction—12 by 17 in. in cross section, and 23 in. in height—permits its utilization of underbench and balcony space not available to larger welding units. Practically averaging one-half pound per ampere of maximum output the 6WK20H weighs 154 lb. and has a current range from 30 to 250 amp., and accommodates electrodes from 1/16 to 5/16 in. diameter. Due



CUTS PRODUCTION AND MAINTENANCE COSTS MORE THAN HALF

SPEEDY LAYOUT JOBS ON PIPE AND STRUCTURAL STEEL

NOW! NEW EASE SPEED WITH THE CONTOUR MARKER



Compact, fits into hip pocket. Efficient, accurate, easy to use. Any of the angles in the illustration to the right can be marked off in five minutes or less on any pipe from 1½ to 18 inches. Save time, labor and gas—will pay for itself in one day.

IF YOUR LOCAL DISTRIBUTOR CANNOT SUPPLY YOU CAN ORDER DIRECT.

TO CONTOUR MARKER CORP.
1843 E. Compton Blvd.
Compton, Calif.

Please send me full details on the Contour Marker

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to the compactness and balanced design a man can readily lift it and put it in the trunk of his car for transporting.

Instant arc striking without any manual adjustment is provided by "Hot Start" automatic control. The correct amount of boost is always furnished for any specific current setting. The ampere range is covered by three overlapping current ranges which permit precise current control. Thus with "Hot Start" control and overlapping current ranges, lower operating costs are assured because idling current is reduced to a minimum and more efficient use of electric power is accomplished.

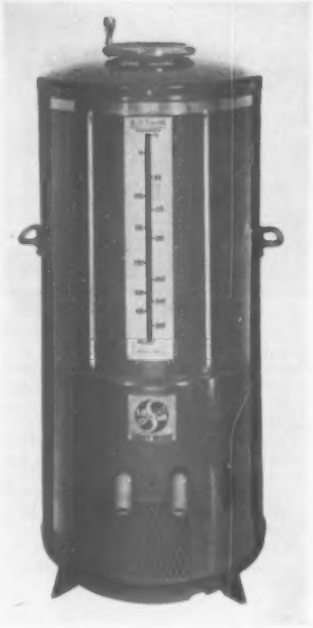
The standard Model 6WK20H contains a primary switch and is designed for operation on 220 v. Models are available for operation on 440 v. with or without power factor capacitors.

Welding Transformer

A new 400-amp. welding transformer—the Model 400E—has gone into production at A. O. Smith Corp. of Milwaukee. It combines new features of low initial cost and easy maintenance and operation, with a welding range of 70 to 585 amp.

Rated in accordance with N.E.M.A. specifications, 400 amp. at 60% duty cycle, it is the first welder in the low cost field to retain the full 75 v. open circuit voltage. With its exceptional range the 400E covers the full heavy duty welder range at an initial price of $395.

Fan motor and capacitors of the 400E are easily accessible for maintenance



through the hinged door at the front of the machine. An industrial type heavy duty motor and fan blade provide high-velocity air to clean and cool the coils. Commercially available secondary connectors are standard equipment.

This welder operates at the usual high efficiencies attributed to transformer type machines providing for important power savings over all types of d.-c. welder installations.

Other features include primary connections mounted in a standard knock-out box for simple installation, and a patented ball bearing coil-moving screw for quiet operation and easy adjustment.

Universal Welders Clamp, Protractor Equipped

Recently announced, the new Bernard Model "B" Pro-clamp incorporates all the features of the original Model "A" plus improved modifications and the advantages of a jaw capacity which has been increased to a full $2^1/_2$ in.

The Pro-clamp provides an inexpensive mechanical means of quickly and accurately positioning and holding plates, rods, bars, angles, tubes, etc., at any angle or degree. It replaces the former method wherein one man held the pieces in position with the aid of a template or protractor while another man did the welding. The savings resulting from reducing a two-man operation to a one-man job are obvious, as are the advantages of the more exact positioning effected.

One face of each of the two rotating clamps has been provided with a flat surface for gripping plates or other flat sur-

The Answer to Many
Welding Repair, Maintenance and Production Problems

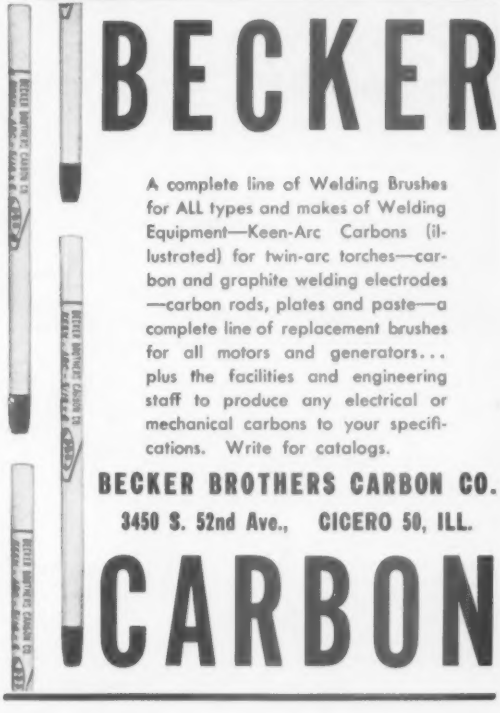
STACKPOLE WELDING PASTE
Can be molded with ease and speed. Quickly torch-dried. Does not blister, spall or crack. 1, 2½ and 5 lb. containers. Prices on request.

Stackpole Carbon Welding Plates and Rounds offer a quick, easy means of restricting metal flow beyond the weld area. They can be used in their original form or easily machined—even by pocket knife—to almost any shape to back up almost any arc or gas method weld. Since they are not wetted by the molten metal, they eliminate sticking or "freezing" to either the parent or fill metal.

Both Plates and Rounds are of a highly refractory carbon material, mechanically strong and capable of withstanding extremely high temperatures. Prices on request. Please state quantity requirements.

STACKPOLE CARBON CO., ST. MARYS, PA.

STACKPOLE
WELDING PLATES AND RODS



BECKER

A complete line of Welding Brushes for ALL types and makes of Welding Equipment—Keen-Arc Carbons (illustrated) for twin-arc torches—carbon and graphite welding electrodes—carbon rods, plates and paste—a complete line of replacement brushes for all motors and generators... plus the facilities and engineering staff to produce any electrical or mechanical carbons to your specifications. Write for catalogs.

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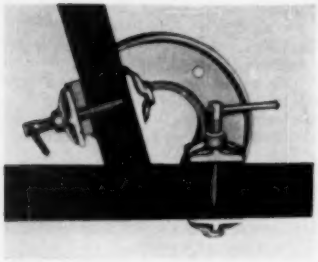
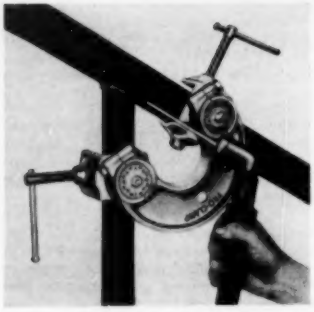
CARBON





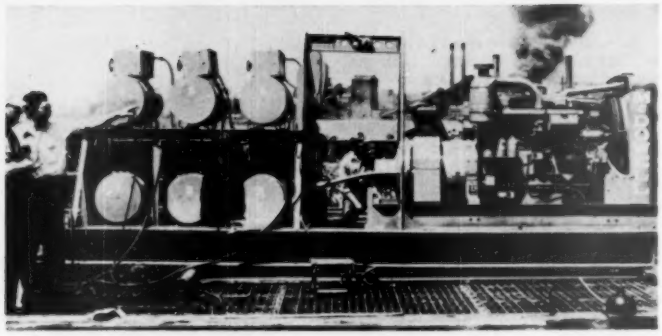
faces; the opposite face of each clamp has a "V" form for handling round or oval pieces. Pressure screws may be shifted from one side of these clamps to the other so that either the "V" or the "flat" side may be placed next to the piece being handled.

Clamps are rotatable in the "U" frame through a full 360°; each clamp has a double 180° machine-cut protractor scale as well as a positive locking device so that the exact angle desired can be easily set and then rigidly held.

Pro-clamps are of all steel construction, cadmium plated to resist weld spatter and retard rusting, and weigh 11½ lb. The ½-in. pressure screws have hardened tips to resist wear. Bulletin No. PBC-1 gives complete data on these new Pro-clamps and may be had for the asking from the manufacturer, Bernard Welding Equipment Co., 741-E. 71st St., Chicago 19, Ill.

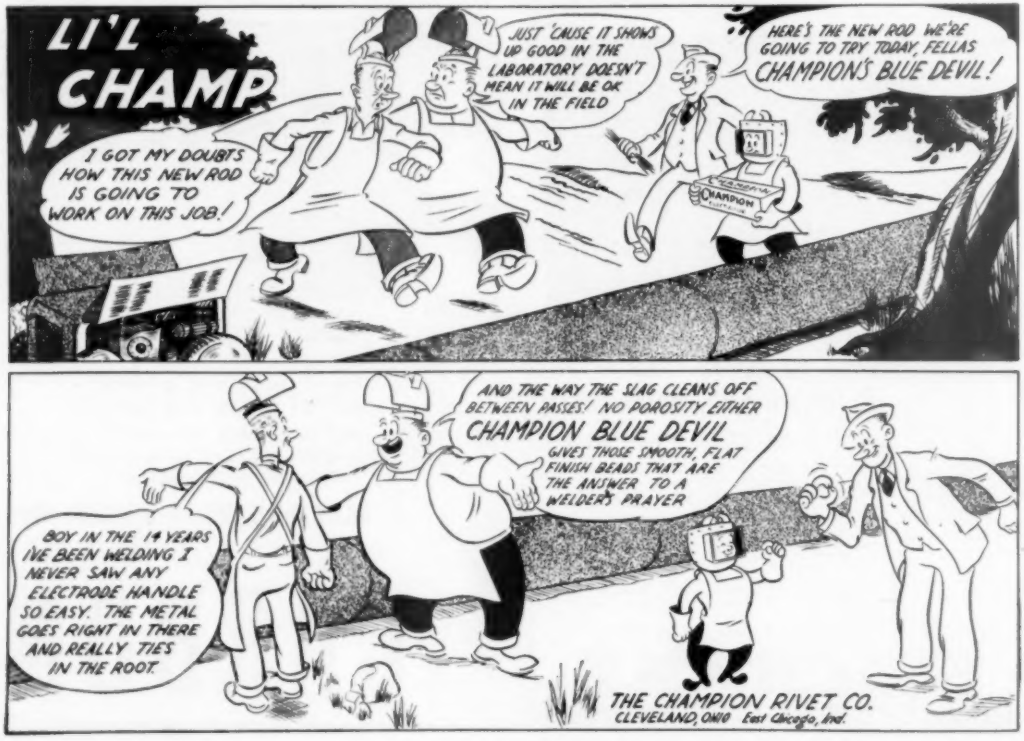
Power Plant for Arc Welders

The McDowell Co., Inc. of Cleveland, Ohio, contractors and engineers, have designed and built an unusually economical source of power for electric drive arc welders, McDowell has constructed two of the units to use on their construction jobs where regular power installations are not practical.



The compact unit is comprised of a Diesel-driven generator and six standard, motor generator, d.-c. welding sets as manufactured by The Lincoln Electric Co. of Cleveland, Ohio. The welders and power unit are built into a special sled which is readily transported to the job and which can be rolled along on pipe rollers, if it is necessary, as the job progresses.

The UD24 International Diesel engine drives a 100-kw. generator which produces 220 or 440 v., 3 phase, 60 cycle power. The generator has a capacity to operate six 300-amp. welders simultaneously


LI'L CHAMP
I GOT MY DOUBTS HOW THIS NEW ROD IS GOING TO WORK ON THIS JOB!
JUST 'CAUSE IT SHOWS UP GOOD IN THE LABORATORY DOESN'T MEAN IT WILL BE OK IN THE FIELD
HERE'S THE NEW ROD WE'RE GOING TO TRY TODAY, FELLAS CHAMPION'S BLUE DEVIL!
CHAMPION
BOY IN THE 14 YEARS I'VE BEEN WELDING I NEVER SAW ANY ELECTRODE HANDLE SO EASY. THE METAL GOES RIGHT IN THERE AND REALLY TIES IN THE ROOT.
AND THE WAY THE SLAG CLEANS OFF BETWEEN PASSES! NO POROSITY EITHER
CHAMPION BLUE DEVIL
GIVES THOSE SMOOTH, FLAT FINISH BEADS THAT ARE THE ANSWER TO A WELDER'S PRAYER
THE CHAMPION RIVET CO.
CLEVELAND, OHIO East Chicago, Ind.

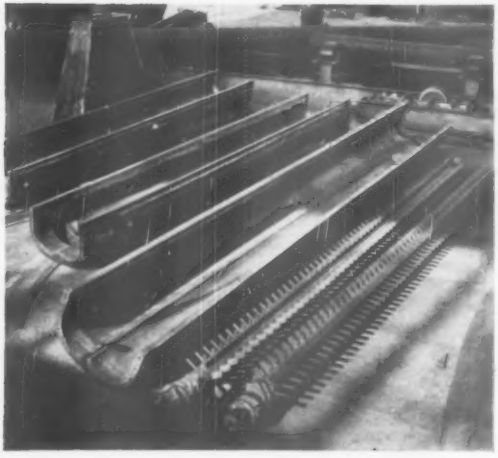
good reasons

to build it of **EVERDUR**

to weld it with **EVERDUR**

Fabricating this welded water-jacketed conveyor-mixer was a tricky job . . . and Everdur* is part of the trick. The material shown is Everdur—sheet, rod, tube, pipe and bar—with the exception of the outer jackets. That's because the job called for a tough, high-strength, non-rust material that would dependably resist corrosion and erosion . . . and that could be readily fusion-welded with a rod of similar characteristics.

Joining those hundreds of Everdur fins to Everdur tubes to form the propeller units in perfect alignment was a tricky job of inert-gas-shielded-arc welding with Everdur rod.



Whether it's a matter of production-brazing, or a complicated repair in a fractured casting, you can usually save time and money with ANACONDA Welding Rods. Our technical men know welding "forwards and backwards." Their advice is yours for the asking.

Write now for Publication B-13, "ANACONDA Welding Rods and Procedures."

ANACONDA Welding Rods are available from distributors throughout the United States. The American Brass Company, Waterbury 20, Connecticut. In Canada: Anaconda American Brass Ltd., New Toronto, Ontario.

*Reg. U. S. Pat. Off.

You can depend on ANACONDA® WELDING RODS

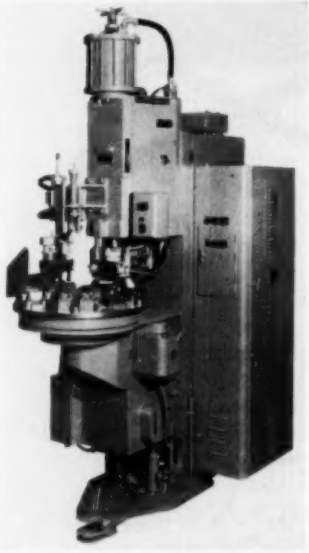
with adequate reserve to run eight welders if desired.

The equipment handling economies of this setup are apparent, and the necessity for a temporary power installation is eliminated. The operation of the unit is also highly economical.

The operation of two such units on a bridge repair job recently done by McDowell proved to be considerably cheaper than operating twelve gas engine welders. The two Diesel units consumed approximately 52 gallons of fuel oil a day, which resulted in a power cost of one cent per kilowatt.

Projection Welder

The Taylor-Winfield resistance-projection welder illustrated is complete with motor-driven geneva indexing table, electrical controls and automatic air-operated "pick-off" work ejector. With no current carrying bearings, it is a dual ram-type welder with five stations and ten dies, welding two assemblies simultaneously.



This machine projection welds a mounting plate of 16-gage low-carbon steel to a telephone part box of 18 gage low-carbon steel. This box contains electrical parts and is considered disposable in case of breakdown. This makes important the increased production of approximately 1600 assemblies per hour as against 400 assemblies per hour on a standard projection welder. Welding labor costs are now only 25% of previous rate with this dial feed projection welder.

Display Boards

These colorful display boards are now made available by Jackson Products, Warren, Mich., to welding and safety supply dealers. One easel board shows the company's line of arc welding electrode holders and cable fittings, while the other shows their goggles and eyeshields. They are offered free with the purchase of the merchandise displayed.

Weld Clamps

A new line of "Jorgensen" Shielded Welders' Clamps has recently been announced. Identified as style no. 120-S, the new line features the distinctive "Jorgensen" Steel Shield over the full length of the clamp screw, to protect the screw in a most positive manner from "spatter" and other damage—even from actual contact with the torch! This feature has been proved most satisfactory and economical in actual service on two lines of "heavy duty" Jorgensen Shielded Welders' Clamps that have been on the market for some years, and is now adapted to a "general service" weight of clamp.

The style no. 120-S "Jorgensen" Shielded Welders' Clamp is available in all sizes from 2 to 12-in. opening capacities; "load limits," depending upon the particular size, run from 2400 to 4500 lb. Free descriptive literature is available upon request to the manufacturers. Adjustable Clamp Co., Dept. WJ, 417 N. Ashland Ave., Chicago 22, Ill.

WELDING CONNECTORS
Saxe System Welded Connection Units
for welded assembly
Saxe Units place in position and securely hold together structural parts to be welded.
As used in many welded structures they eliminate all hole punching producing an economical, rigid, safe and quickly erected structural frame.
"Write for 58 pg. Manual containing full engineering design information for welded structures."
J. H. Williams & Company
Buffalo 7, New York
G. D. Peters Company
Montreal 2, Canada
Canadian Representative

SPECIALIZING IN "BETTER-BUILT"
WELDING & SAFETY EQUIPMENT
FOR OVER 25 YEARS
There is a DOCKSON distributor near you. Name on request.
Write for our complete Welding and Safety catalogs.
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PERSONNEL

Isaac Harter Is Elected a Fellow of the A.S.M.E.

Isaac Harter, chairman of the board of The Babcock & Wilcox Tube Co., has been named a Fellow of The American Society of Mechanical Engineers, it was made public recently. The grade of Fellow, conferred upon engineers with at least 25 years of active practice or teaching experience in a school of accepted standing, with acknowledged engineering attainment, is an honor held by less than 1% of the A.S.M.E. membership.

During his long association with The Babcock & Wilcox Co., Mr. Harter made many significant contributions to industrial progress. One of the most far reaching was his work in conducting experiments leading to the discovery that chemical action of water in boilers under pressure was causing "caustic embrittlement" resulting in cracking of expensive boiler drums. Proper treatment of feed waters then eliminated the hazard.

During the building of Hoover (Boulder) Dam, Babcock & Wilcox were asked to supply the penstocks—mammoth welded tubes of heavy steel varying from 8½ to 30 ft. in diameter and totaling 14,500 ft. in length. No one else had ever fabricated such devices, but Mr. Harter selected a crew of B & W engineers and successfully handled the huge job.

In other phases of the field of welding Mr. Harter was also responsible for broad developments. A good example was his work in applying sound metallurgical principles to the welding of heavy boiler drums, sometimes as much as 4 in. in thickness and never previously successfully welded for commercial use. He also developed the method of applying X-ray to the testing of such welds, eliminating the previous costly methods of destroying ever so many welds for inspection to keep a continuing control on quality.

He was active in the group which helped develop the A.S.M.E. Boiler Code and was instrumental in securing early legislation in Ohio resulting in the Ohio Boiler Code. He helped form the American Uniform Boiler Law Society to secure uniform state laws coordinating the A.S.M.E. code.

Mr. Harter personally directed development of the process of continuous casting of steel at the Beaver Falls, Pa., plant of The Babcock & Wilcox Tube Co. He has been active in producing improved refractory materials and products by developing generalized and economic tonnage methods for melting materials at temperatures above 1800° C. He has worked with the Atomic Energy Commission on its industrial advisory committee and now serves on its patent compensation board. He holds a number of patents for boiler design, industrial furnaces, refractories, continuous casting and steel making machinery.

Mr. Harter graduated from the University of Pennsylvania in 1901 and became associated with the Aultman-Taylor Machine Co. of Mansfield, Ohio. When this company was acquired by the Stirling Co. in 1904 Mr. Harter was placed in charge of the boiler works at Barberton, Ohio. He became associated with The Babcock & Wilcox Co. in 1906 when it purchased the Stirling Boiler Co.

In 1910 Mr. Harter became superintendent of the Bayonne, N. J., works and was made assistant to the president in 1920. In 1924 he was made vice-president and a director of The Babcock & Wilcox Tube Co. In 1947 he was elected chairman of the board of that company and became a consultant to The Babcock & Wilcox Co.

He became an associate member of the A.S.M.E. in 1908 and in 1921 was made a full member.

Elevation to the grade of Fellow is by election of the A.S.M.E. Council, supreme governing body of the Society. Engineers who have been members of the Society for 13 years may be proposed to the Council by five Fellows or members.

In addition to his work with the A.S.M.E. Mr. Harter is a member of the Institute of Metals (London) and Sigma Xi, and was vice-president of the American Welding Society in 1944–45.

Sam Byers Appointed Representative

The Mir-O-Col Alloy Co., 312 N. Ave. 21, Los Angeles, Calif., manufacturers of Hard-Facing Alloys, announces the appointment of Sam Byers of 3134 E. 10th St., Oakland 1, Calif., as representative for part of the Western States which includes Washington, Montana, Utah and part of California and Nevada, as well as the Province of British Columbia and Alberta in Western Canada. Mr. Byers was formerly connected with the Glenn Roberts Mfg. Co., and is well known in that territory.

Kallen Wins Davis Award

The paper by H. P. Kallen of Cornell University entitled "Weldability—Challenge to Research," published in the March 1950 issue of the Cornell Engineer, has been awarded first prize in the 1949–50 A. F. Davis Undergraduate Welding Award Contest.

To stimulate undergraduate interest in welding, funds have been donated to the American Welding Society for the A. F. Davis Undergraduate Welding Award. This award consists of four cash prizes given annually to authors and publications for the best and second best articles on welding published in undergraduate magazines or papers during the preceding year. Any undergraduate of a college or university in the United States or Canada is eligible, but the paper must be published in an undergraduate publication.

The A. F. Davis Undergraduate Welding Award is named for its donor, A. F. Davis, vice-president and secretary of the Lincoln Electric Co., Cleveland, Ohio.

After graduating from the High School of Music and Art, New York City in 1942, Mr. Kallen worked as a mechanical draftsman for the Westinghouse Electric Corp. for a period of eighteen months. He then entered the Army Ordnance Department where, as an enlisted man, he instructed in basic training. After graduating from Ordnance Officer Candidate School, he operated a machine shop for automotive repair, in Italy, for thirteen months.

Upon being separated from the service in 1946, Mr. Kallen entered the Sibley School of Mechanical Engineering at Cornell University, from which he will graduate with the B.M.E. degree this June. In addition to his school work, he has been a member of the Track and Cross-Country teams, Assistant Editor of the Cornell Engineer and Vice-President of Tau Beta Pi. Other honorary societies include Phi Kappa Phi, Pi Tau Sigma, and

the Journalism Society, Pi Delta Epsilon.

Mr. Kallen has accepted a position as a student development engineer for the Hyatt Roller Bearings Division of General Motors. He will also attend evening Graduate School of Mechanical Engineering at Columbia University this September.

Johnston Joins Michigan

The University of Michigan announces the appointment of Dr. Bruce G. Johnston as Professor of Structural Engineering in the Department of Civil Engineering, College of Engineering.

Dr. Johnston comes to the University of Michigan from Lehigh University where he has been a member of the Civil Engineering staff for the past twelve years, serving with the rank of Professor since 1945 and as Director, Fritz Engineering Laboratory since 1947.

Professor Johnston obtained his B.S. in Civil Engineering at the University of Illinois in 1930, his M.S. at Lehigh University in 1934, and his Ph.D. (Civil Engineering) at Columbia University in 1938. He held a research fellowship at Lehigh University from 1932 to 1934 and was an instructor in Civil Engineering at Columbia University from 1934 to 1938.

On leave of absence from Lehigh University 1942–45 he served as a structural engineer, Bureau of Yards and Docks, U. S. Navy 1942–43 and as Senior Engineer and Project Supervisor, Applied Physics Laboratory, Johns Hopkins University 1943–45. He was awarded the James R. Croes medal by the American Society of Civil Engineers in 1937. He is a member of Sigma Xi, Tau Beta Pi and Chi Epsilon. He also holds memberships in the American Society of Civil Engineers, American Society of Mechanical Engineers, American Society of Metals, American Welding Society, Society for Experimental Stress Analysis and the International Association of Bridge and Structural Engineers. He is, at present, serving on the following technical research committees: American Institute of Steel Construction Structural Steel Committee; Column Research Council, *Chairman*, Committee on Research; Welding Research Council (*a*) Structural Steel Committee and (*b*) University Research Committee. He is a member of the Executive Committee of the Graduate Faculty of Lehigh University.

Dr. Johnston will assume his duties on the teaching staff at the University of Michigan in September, after moving his family into their new home at 1310 Beechwood Ave., Ann Arbor.

Morse Honored

E. J. Morse, Executive Vice-President of The Burdett Oxygen Co. was given a testimonial dinner by his associates on Friday, April 28th in honor of his twenty-

five years with the company. In 1925, when Mr. Morse joined Burdett, the firm had only five employees; today it employs over two-hundred people, as well as having branches in Akron, Mansfield, Cincinnati and Los Angeles, Calif. William Loveman, President of Burdett, says: "This is a tribute that Gene well deserves, for his work and loyalty to his fellow employees. He has been instrumental in establishing many of the policies which have helped us grow in size but still remain a coordinated family of workers." Burdett has received recognition many times for its employee-relation activities and was honored during the war for its outstanding service to the armed forces.

Changes in Lincoln Field Engineering Staff

A. F. Boucher has assumed sales and engineering responsibilities as District Manager in the Detroit area for the Lincoln Electric Co. of Cleveland, Ohio. Mr. Boucher has been transferred to Detroit from Milwaukee where, since 1947, he has been District Manager. Boucher, however, is no newcomer to Detroit. He served this area as District Manager prior to his assigment in 1942 to the Ordnance Department of the Army. As an officer in this department, he served in Detroit as a

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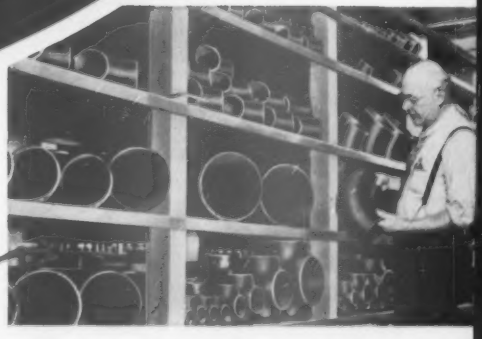

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member of the Ferrous Metallurgical Advisory Board on the subcommittee on welding rolled and cast armor. He helped develop important new welding techniques which improved and speeded the fabrication of tanks and armored vehicles. Boucher joined the Lincoln staff in 1936 after his graduation from the Ohio State University and has been continuously engaged in sales and engineering.

I. R. Bartter has been transferred from Grand Rapids to Milwaukee to assume the responsibilities of District Manager in that area for the Lincoln Electric Co. Bartter has been District Manager for Lincoln in Grand Rapids since 1945. He joined Lincoln in 1936 and has been active in the development of welding applications for automatic and semiautomatic welding. Bartter served the Duluth, Minn. territory for approximately eight years where, during the war, he was helpful in expanding the important shipbuilding activities in that area.

H. W. Beck Promoted

Henry W. Beck has been appointed advertising manager of Airco Co. International, it was disclosed recently. He was formerly assistant manager, advertising department for Air Reduction Sales Co. Both companies are divisions of Air Reduction Co., Inc.

Mr. Beck has been with the Air Reduction organization for over twenty years. After serving in a number of capacities, he was made assistant manager in 1945, a position he held until his recent appointment.

His functions with Air Reduction Sales Company, according to G. Van Alstyne, advertising manager, will be divided between William C. Bettes and George M. Worden. Mr. Bettes will be responsible for trade paper advertising, Mr. Worden for publicity.

Mr. Beck will be located at the general office of Airco Co. International, 33 W. 42nd St., New York 18, N. Y.

Templeton Promoted

J. E. (Earl) Templeton, Manager of the Los Angeles branch of P. R. Mallory & Co. Inc., Indianapolis 6, Ind., is succeeding Walter Harvey as manager of the company's Wholesale Division.

Resigning because of his desire to do less traveling and to live in a milder climate, Mr. Harvey remained active with the company until after the May trade show in Chicago.

Mr. Templeton has been with Mallory for a number of years, having spent the major portion of this time in the Wholesale Division in several capacities. Prior to his Los Angeles assignment, he had been associated with Mr. Harvey as sales service manager of the Wholesale Division.

Mr. Templeton is being succeeded in Los Angeles by Charles Gutheil, another Mallory veteran, who has been sales manager of the company's Switch Division.

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Users report Lincoln's new "LJ-1" as tops in electrode holders today. Unique "lock-jaw" construction gives extra strong grip, yet easy thumb release. Compact for welding in close quarters without danger of grounding. Jaw end is covered by heat-resistant cover that cannot be broken. Rated 250 amps. Handles 1/16" to 3/16" diameters.

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Lincoln insulated holders sell at amazingly low prices . . . *actually less than often charged for replacement parts on other makes of holders.*

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"Stable-Arc". The industry's top standard of quality. Unsurpassed for long wear . . . for resistance to impact. Has braided cotton reinforcing to impart added strength to its molded rubber casings.

HIGH QUALITY WELDING LENSES

"Hi-Visibility". Lincoln's new, high quality, low priced welding lens. Surpasses Federal specifications for filtering out undesirable arc rays.

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LINCOLN

SECTION ACTIVITIES

prepared by C. M. O'Leary

Anthony Wayne

The May Meeting held on the 25th was the party meeting at which the wives as well as the members participated. The party was held at the Orchard Ridge Country Club. Cocktails were served from 6:00 to 7:00 P.M.; dinner from 7:00 to 8:00; business meeting from 8:00 to 8:30, and the balance of the evening was spent in round and square dancing and other pleasures. An attendance of 96 persons made this party a huge success.

The following officers were elected for the 1950–51 season: *Chairman*—Harold Hamilton; *1st Vice-Chairman* — Harry M. Johnson; *2nd Vice-Chairman*—R. F. McNutt; *Secretary*—Kenneth H. Zimmerman; *Treasurer*—Gerald E. Springer; *Technical Representative*—Jacob S. Geiser and *Chairman Membership Committee*—George H. Laws.

The following Board members were elected: *two year term:* George A. Leaming, Harold C. Smith, Fred C. Wilt, Robert C. McCoy, Lorin A. Snyder and I. A. Rollins; *one year term:* Garold Boggs, Max L. Clark, D. B. Rice and Niels S. Hansen.

Arizona

The annual meeting of the Arizona Section was held in the Colonial Room of the Hotel Westward Ho at 7:30 P.M. on June 7th. It was a dinner meeting of nontechnical nature and there were approximately 45 persons in attendance.

As a result of the annual election the following officers and executive committee were elected: *Chairman*—James A. Johnson; *1st Vice-Chairman*—Arthur W. Tesmer; *2nd Vice-Chairman*—J. August Rau; *Secretary*—Walter E. Riley; *Treasurer*—F. Morris Aspey and *Executive Committee*—A. W. Tesmer Gustav Waago, Gil Dye, W. F. Fischer, Hal Savage, G. D. Hunt, R. H. Smith and R. H. Hoffman.

The Section was fortunate in having three fine speakers who are the state winners of the annual speaking contest conducted by the Toastmasters Club. They were Albert Pate, Charles Cammarata and Jesse McClain. They gave very excellent talks and were graphic examples of the best in public speaking. Following the talks a very interesting sound film was shown through courtesy of the Valley National Bank.

Chattanooga

The following officers have been elected for the coming year by the Chattanooga Section: *Chairman*—W. L. Herbst; *Vice-Chairman*—George L. Tepley; *Secretary-Treasurer*—C. T. Ward; *Members at large* (two years) Fain Ingram and Roy Mitcham; *Members at large* (one year) J. W. Kelker, L. F. Pohl and C. E. Phillips.

Colorado

The April 11th dinner meeting was held in the Fiesta Room of the Oxford Hotel, Denver, Colo. J. E. Dato, District Service Supervisor for The Linde Air Products Co., gave a very interesting illustrated talk on "Developments and Improvements on Oxyacetylene Cutting Procedures" which was well received by a fairly large audience.

The June meeting was held on the 6th at Silver Wing Inn. Dinner speaker was H. Jackson of the Tompson Pipe & Steel Co., who spoke on what the A.W.S. should mean to every firm and individual interested in welding. A Bureau of Mines film "Wonderful Wyoming and Its Natural Resources" was well received. There was no technical speaker at this meeting. Mr. Jacobi, the incoming chairman, was presented together with H. B. Klodt, Vice-Chairman, who made some appropriate remarks.

Columbus

Williams & Co. were host to the Columbus Section on Friday, May 19th, for the final meeting of the season.

Following a buffet supper, Fred Plummer, District Vice-President, presented Walter Hanes, retiring chairman, with a past-president's pin. After this event, Mr. Plummer introduced the following officers for the coming year: *Chairman*—Prof. R. S. Green, Ohio State University; *Vice-Chairman*—Carl J. Alff, Clark Grave Vault Co.; *Secretary*—Melvin C. Clapp, Battelle Memorial Institute; *Treasurer*—John Purdy, Buckeye Steel Castings Co.; *Technical Representative*—R. J. Krieger, Ohio State University and *Directors* (for one year) Howard Cary, Perry Rieppel, John Snodgrass; (for two years) R. H. Rishell, H. S. Hirsh and R. R. Grant.

At the conclusion of installation of officers an entertainment program was put on, which by the remarks overheard, was a howling success.

Dallas

The May meeting was held on the 24th in the Auditorium of the Lone Star Gas Co. Dinner was held in Brockles Restaurant.

Speakers at the technical session were Schuyler A. Herres and Dr. Charles Moore, who spoke on Metallurgy of Titanium. New facts and developments were presented.

An entertaining sound color film on Tarpon fishing was shown following the technical session. Refreshments were also served.

The following officers were elected by the Section to serve for the 1950–51 season: *Chairman*—John A. Wilson; *1st Vice-Chairman*—J. J. Brown; *2nd Vice-Chairman*—R. W. McClain and *Secretary-Treasurer*—John P. Viglini.

Hartford

On Thursday, May 11th the Hartford Section met for dinner and meeting in the Rockledge Country Club, West Hartford, Conn.

Speaker at the technical session was G. C. Kiefer, Associate Director of Research, Allegheny Ludlum Steel Corp. Mr. Kiefer spoke on Welding Stainless Steels, presenting new facts and developments which were interesting.

Houston

The May 25th meeting was held at the Ben Milam Hotel, Houston, Tex. Dr. Charles H. Moore and Schuyler A. Herres presented an interesting address on the Metallurgy of Titanium.

Indiana

The April 28th dinner meeting was held in Buckley's Restaurant, Cumberland, Ind. Frank Pollard of Woodall Welding Co., Indianapolis (formerly Master Mechanic of National Malleable & Steel Castings Co.) gave a clear-cut talk on Maintenance Welding. Slides were used to illustrate the talk.

The following officers have been elected for the 1950–51 season: *Chairman*—Karl Kick; *1st Vice-Chairman*—Arthur F. Wood; *2nd Vice-Chairman*—Alexander B. Pollock; *Secretary*—Mrs. Mercedes Spotts; *Treasurer*—Paul F. Grubbs; *Chairman Membership Committee*—A. B. Pollock and *Technical Representative*—James Salatin.

The May 26th dinner meeting was held in the plant of the Midwest Tank and Construction Co., Indianapolis. Technical speaker was William Tawse of the Nelson Stud Gun Co., Lorain, Ohio. Mr. Tawse' talk on Stud Gun Operation was illustrated with motion picture and demonstration.

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Long Beach

The following officers were elected for the 1950–51 season: *Chairman*—M. K. McEniry; *Vice-Chairman*—M. A. Teeter; *Secretary-Treasurer*—C. L. Breese; *Chairman, Membership Committee*—Paul Anderson; *Chairman, Program Committee*—M. A. Teeter; *Publicity Chairman*—Tommy Lewis; *Advertising Chairman*—O. B. Bowers; *Codes and Technical Chairman*—Ray L. Davenport; *Ways & Means Chairman*—Don Carlson; *Entertainment Chairman*—Stuart Jewell and *Directors*—R. G. Beckstrand, Lee Schexschnider, Paul De Pietro, O. B. Bowers, Ray L. Davenport, Paul E. Anderson, A. J. Smoak and A. H. Butler.

Los Angeles

The Los Angeles Section held its regular monthly dinner meeting on May 18th. The speaker was M. M. Griffith, Welding Engineer, Southwestern Engineering Co., Los Angeles. His subject was "The Aircomatic Process as Related to Alloy Fabrication."

The speaker first described by the aid of simplified diagrams the equipment and how it operated. Photographic slides of his own mobile unit adaptation were also shown to further assist in visualization of the equipment. Discussion of welding procedures consisted of sketches showing the welding grooves and scarfs followed by photographs of cross sections of welds made by applying the procedures. The speaker described "on-the-job" operation of the equipment and showed slides of its use in his shop. It was his opinion that the process was best adaptable to repetitive jobs. The equipment showed two extremely advantageous features in its extreme mobility and versatility in welding in all positions. Other features consisted of high-deposition rate, high-current density, rapid burn-off rate and absence of flux to clean after welding. The metals welded by the process consisted of aluminum, bronze and brass and chrome-nickel steels.

Mr. Griffith displayed a thorough knowledge of his subject in answering numerous questions which were requested at the conclusion of his talk.

Louisville

The following officers were elected for the year 1950–51 at the dinner and business meeting in the Preston Kunz Restaurant on May 16th: *Chairman*—Edmund F. Schulz; *Vice-Chairman*—Clarence E. Schiller; *Secretary*—Arthur B. Doudna; *Treasurer*—Carl L. Jensen; *Directors*—H. C. Powers, J. X. Merkt, L. G. Rickert, E. H. Dilley and Roba Yarbro.

Michiana

The Michiana Section finished the season with a business and social meeting held in the Bendix Legion Hall, Thursday, May 18th. The annual election put the following men in office for the 1950–51 season: *Chairman*—A. B. White; *Vice-Chairman*—George Mittler; *Secretary-Treasurer*—W. G. Fassnacht; *Executive Committee* (1 year) H. Biederstaedt, F. H. Craven and M. A. Miller; (2 years) H. J. Hecht, Donald Kyler and George Fetherston.

Following the business meeting there was shown a display of types of welding done in, of all places, a "pill factory," the Miles Laboratories of Alka-Seltzer fame.

Then came the highlight of the evening, a bingo party with lunch and refreshments, the object of the bingo being some thirty prizes donated by welding supply houses in the area.

New York

The New York Section held its Annual Smoker at the Building Trades Employees Club on May 12th. Approximately 300 members and their guests attended the affair, which was arranged by George Schneider and his Committee.

At this meeting, the results of the election of officers for the 1950–51 season were announced as follows: *Chairman*—G. L. Baker; *1st Vice-Chairman*—H. R. Clauser; *2nd Vice-Chairman*—A. M. Setapen; *Secretary-Treasurer*—S. T. Walter and *Members at Large* (two year term) J. E. Kneiper; (three year terms) J. T. Stewart, W. L. Mitzenius, W. B. Brower and D. Swan.

On April 18th, the New York Section participated in the "Battle of the Sections" meeting of the New Jersey Section. The New York team was composed of the following members: R. S. Donald, A. N. Kugler and Jay Bland. A. M. Setapen officiated as one of the judges, and J. Lyell Wilson served again admirably as moderator. Both are members of the New York Section. The other competing teams were from the Bridgeport, Philadelphia and New Jersey Sections. The New York Section team was judged the winner of this closely contested and interesting competition.

Niagara Frontier

Annual Welding Symposium under the auspices of the Niagara Frontier Section of the American Welding Society and the School of Engineering, University of Buffalo, took place on April 21st and 22nd at the University of Buffalo. Design for Welding was the theme of the Symposium.

The following outstanding speakers and papers were presented: "Selection of the Best Welding Process," Walter Mehl, Welding Engineer, Heintz Mfg. Co.; "Introduction to Metallurgy of Welding," Dr. Robert D. Stout, Professor Metallurgy Lehigh University; "Welding Applications in Automotive Construction," John F. Randall, Supervisor of Welding Development, Ford Motor Co.; "Control and Inspection of Alloy Welding," N. C. Jensen, Welding Engineer, Babcock & Wilcox; "Design for Welding of Aluminum," G. O. Hoglund, Welding Engineer, Aluminum Company of America; "Design for Resistance Welding," W. R. Plummer, General Sales Mgr., Progressive Welder Co. and "Welded Structural Design," LaMotte Grover, Structural Welding Engineer, Air Reduction Sales Co.

A very fine live exhibition of various welding and cutting processes accompanied the sessions, together with exhibits of the products made by local manufacturers using welding.

Northern New York

The following Section officers were elected to serve for the 1950–51 season: *Chairman*—J. P. Frandsen; *Secretary-Treasurer*—I. W. Johnson; *1st Vice-Chairman*—A. G. Craig; *2nd Vice-Chairman*—S. A. Herres; *Chairman Membership Committee*—D. S. Packard; *Chairman Program Committee*—R. M. Curran and *Technical Representative*—C. M. Rhoades, Jr.

Oklahoma City

The regular monthly dinner meeting of the Oklahoma City Section was held on May 23rd at the Biltmore Hotel. Speaker at the technical session was S. A. Herres of the Allegheny Ludlum Steel Co. Since Titanium Ores have been a by-product of the lead industry, Mr. Herres was accompanied by Dr. Charles Moore, who is associated with The National Lead Co.

Dr. Moore brought a very interesting lecture on Titanium from the standpoint of mining, partial refining and its many valuable uses. Also, just as a sideline he put on display a large collection of gems which are made of rutile which is a titanium dioxide mineral.

Mr. Herres presented a paper on the Physical and Chemical Properties of Titanium. The paper was presented in such a way that it was well received by everyone in attendance.

Pascagoula

A business dinner meeting was held on Tuesday evening, May 23rd in the Ingalls Shipyard Cafeteria. Dinner speaker was W. D. Pelan who spoke on the Tax Structure of Pascagoula. As this was a business meeting there was no technical speaker.

Philadelphia

The First Annual Dinner Dance of the Philadelphia Section was held on May 13th and was a decided success. Seventy couples enjoyed the delicious dinner served by the Engineers Club and the dancing and sociability following. Each lady was presented with a corsage in addition to the fine door prizes won by many of the ladies.

The enthusiasm and success of this venture has established it as an annual part of the Philadelphia Section Program.

The Annual Executive Committee Meeting was held June 7th at the Rolling Green Golf Club. The plans and progress of the Philadelphia Section for the year 1950–51 were discussed. The outlined arrangements assure a fine technical and all around program that can bring

great benefits to the members of the Section.

R. A. Guenzel, Southern Oxygen Co. and C. W. Middlestead, C. W. Middlestead Co. were appointed to fill the vacancies for two unexpired terms on the Executive Committee.

The following assignments were made as Committee Chairmen: *Program Committee*—K. W. Ostrom; (*a*) subcommittee on Publicity and Meeting Notices—M. J. Bonowitz and (*b*) subcommittee on Procurement of Advertising Sales—L. D. T. Berg. *Committee on Report to Headquarters*—R. D. Bradway. *Discussion Group Committee*—J. Erlacher. *Special Activities*; (*a*) Battle Within the Section—A. J. Raymo and (*b*) social activities—R. A. Guenzel. *Membership Committee*—C. W. Middlestead.

Rochester

On April 17th dinner was served to the Rochester Section at Guy Michael's Restaurant. The business meeting at the University of Rochester began with a movie on "Copper Mining, Smelting, Refining," which showed copper from the mine to the finished product. E. W. Brown, Sales Engineer, American Brass Co., spoke on braze welding for production economy. He pointed out that brazing proved to be the cheapest production procedure due to the fact that cleaning and finishing costs were kept at a minimum in most cases. He discussed the possibilities and growing use of new nickel silver brazing alloys for the repair and production fields. The discussion which followed Mr. Brown's lecture brought out some very interesting production problems and their solution.

The Rochester Section held their annual election and installation of officers on May 15th at the Century Sweet Shop. The following officers were elected: *Chairman*—Herbert Fetter; *Vice-Chairman*—Harry Stoler; *Secretary-Treasurer*—Walter Dick and *Executive Committee*—Edward Habel, Gomer Stelljes, Robert Jackson, Fred Hall, Gordon Kane, Harvey Ainsworth and Clyde Martin.

After the election, entertainment was furnished by a barber shop quartet from Dansville, N. Y. This was followed by the usual good food and other refreshments served at these annual meetings.

Saginaw Valley

The regular monthly dinner meeting of the Saginaw Valley Section for April was held on the 6th at the Chanticleer, Saginaw, Mich. New officers for 1950–51 were introduced as follows: *Chairman*—Irving C. Mattson; *Vice-Chairman*—Joseph E. Neuman; *Secretary*—Morris D. Thomas; *Treasurer*—John R. Clayton; *Executive Committee* (1 year) Paul Klain, Joseph Stanulus, George Blain and Dean Knight; (2 years) Godfrey Burrows, Russell E. Ferris, Edward Leach and Lloyd Heidelbright and *Past Chairman*—Raymond Ladd.

George Enk and Ed Reed of Weltronic Co., Detroit, presented a Resistance Welding Control demonstration which includes a working schematic diagram of a welding control.

At the request of the Section, Mr Smith of the Detroit office of Handy and Harman showed "Fundamentals of Silver Alloy Brazing," a thirty minute sound color film.

The May 26th meeting (Ladies Night) was held at Zehnders, Frankenmuth, Mich. Charles Albee Howe of the Chicago Tube and Iron Co. gave a very excellent presentation of his 9000 mile tour of the West. This was a series of color slides made by Mr. Howe and includes many of his prize winners. This illustrated the talk which was about 1 hr. long. Taos, Painted Desert, Petrified Forest, Grand Canyon, Bryce, Zion, Sequoia, Yosemite and many others were shown in the slides.

Western Massachusetts

On Tuesday, May 9th at Blake's Restaurant in Springfield, the Western Massachusetts Section heard W. T. DeLong, Research Engineer with the McKay Co. of Pittsburgh, Pa. speak on "Arc Welding Stainless Steels." Mr. DeLong pointed out the principal objectives in the arc welding of stainless steels as follows: 1—to obtain strength and ductility in the joint, 2—to achieve and maintain corrosion resistance, 3—to achieve a welding operation which from the standpoint of mechanics and control is good. Following Mr. DeLong's talk there was a short but interesting question and answer period.

Wichita

A guided tour of the Socony Vacuum Refinery plant at Augusta took place on May 5th. An exceptionally good attendance enjoyed the tour and dinner which followed at Lehr's Cafe. The entire event was a definite success and was enthusiastically received. Meeting night was changed from Monday to Friday to permit attendance by the new Kansas State Chapter student members. The Kansas State Chapter deserves commendation as 11 of their members drove approximately 130 miles by auto to take part in this program.

Technical speaker was H. C. Bates, Master Mechanic, Socony Vacuum Refinery, who spoke on Refinery Construction and Maintenance by Welding. Mr. Bates very comprehensively explained how the development of welding has enabled the oil refinery industry to develop from the old batch process into the modern continuous process method. This gradual conversion was quite interestingly traced in the plant visited earlier by the group.

Worcester

The following Section officers were elected to serve for the 1950–51 season: *Chairman*—C. Myron Burnett; *Secretary*—Wesley A. Proctor; *Treasurer*—John F. Sloan; *1st Vice-Chairman*—William Walker and *2nd Vice-Chairman*—Arthur W. Knight.

abstracts of CURRENT WELDING PATENTS

Prepared by V. L. Oldham

Printed copies of patents may be obtained for 25¢ from the Commission of Patents, Washington 10, D.C.

2,504,867—Method of Arc Welding—Albert Muller, Brooklyn, N. Y., assignor to Air Reduction Co., Inc., New York, N. Y., a corporation of New York.

Muller's arc-welding method comprises the novel step of feeding a gaseous mixture of the arc from at least one source extraneous from the arc so that the gases around the arc comprise about 10 to 50% of carbon dioxide and 50 to 90% carbon monoxide.

2,504,868—Electric Arc Welding—Albert Muller, Plainfield, Glenn J. Gibson, Somerville, and Nelson E. Anderson, Scotch Plains, N. J., assignors to Air Reduction Co., Inc., New York, N. Y., a corporation of New York.

This arc-welding method relates to the use of an aluminum electrode which has argon fed to the arc to shield same. The electrode is fed to the arc at least at 100 in. per minute. Also the current density varies within a given range dependent upon the rate of feed of the electrode in inches per minute.

2,504,921—Worm Welding Device—Haskell C. Carter, Portland, Ore., assignor to Iron Fireman Mfg. Co., Portland, Ore.

This patent covers a special welding device for use in welding the spiral flight to the core of a conveyor worm.

2,505,049—Electric Powder Control—Arthur M. Keller, Livingston, New Jersey, assignor to The Linde Air Products Co., a corporation of Ohio.

Keller's patent relates to a magnetic powder feed system wherein gas under pressure urges a magnetic powder to flow through a conduit. An adjustable magnetic field producing means is operatively associated with the conduit to control the quantity of powder flowing in a stream in the conduit dependent upon the strength of the magnetic field.

2,505,054—Inert Gas Blanketed Oscillatory Welding Rod Feed—Thomas McElrath, Jr., Chatham, and Francis M. Drake, Cranford, N. J., assignors to The Linde Air Products Co., a corporation of Ohio.

A welding method including the steps of maintaining an arc with respect to the work so as to form a weld puddle is disclosed in this patent. An annular stream of non-oxidizing gas is supplied to the arc and puddle to blanket same and the arc and filler rod are simultaneously oscillated with relation to the weld puddle. The filler rod is advanced into the weld puddle during the oscillation of the filler rod.

2,505,077—Welding Rod—William A. Wissler, Niagara Falls, N. Y., assignor, by mesne assignments, to Union Carbide and Carbon Corp., a corporation of New York.

The patented welding rod includes at least one of the group of metals consisting of molybdenum and tungsten with the amounts of such material present in the rod having certain maximum ratios with relation to each other. Carbon, silicon, manganese, chromium, phosphorus, boron and iron complete the weld rod.

2,505,291—Welding Flux—Ernest H. Lucas and John J. Shea, Worcester, Mass.

This novel welding flux comprises 40 to 85% of sodium tetraborate, 10 to 40% of neutral sodium fluoride and 5 to 20% of sodium hydroxide.

2,505,640—Screen for Welders, Grinders or Furnacemen—Stanely Fitch, Woodford Green, England, assignor to Asea Electric Limited, London, England, a company of Great Britain.

A special type of a screen for use by welders and other workers is covered in this patent and it includes a body harness for mounting the screen on the worker.

2,505,937—Welding Electrode—Arthur A. Bernard, Chicago, Ill., assignor to National Cylinder Gas Co., a corporation of Delaware.

This patented electrode is particularly adapted for use in arc-welding operation and includes a metal core having longitudinally extending ribs on the periphery thereof. A sheath of flux material surrounds the core and occupies the grooves formed between the ribs and overlies the ribs. At least one spiral conductor is positioned on the core and embedded in the sheath to conduct current from the surface of the electrode to the core.

2,506,277—Welding Gun—Joseph J. Panik and William T. Wilbur, Detroit, Mich., assignors to United States Engineering Co., Detroit, Mich., a firm composed of Joseph J. Panik.

A fluid actuated welding unit is disclosed in this patent and it has a cylinder with a piston reciprocably positioned in the cylinder and with a piston rod extending from the cylinder and secured to the piston. An electrode is mounted on the rod and means are provided for attaching the unit to a support. These attaching means include a universal joint and a fastening member to permit adjusting movement of the unit and securing of the unit in an adjusted position.

2,507,022—Resistance Electric Welding Method—Jacques Languepin, Paris, France, assignor to Societe La Soudure Electrique, Paris, France, a corporation of France.

This electric resistance welding method utilizes one pair or electric condensers each of which is adapted to be connected to a supply transformer through a related switch and to a welding transformer. The method comprises sequentially charging the condensers of each pair in reverse direction, and successively discharging the condensers of each pair into the primary winding of the welding transformer through a contact of a related switch.

2,507,751—Welding Flux—Alvin L. Bennett, National City, Calif., assignor to Solar Aircraft Co., San Diego, Calif., a corporation of California.

This fluxing powder disclosed by Bennett's patent includes calcium fluoride, sodium fluoride, lithium fluoride, borax glass and boron oxide.

2,508,103—Condenser Welding System—John W. Dawson, Auburndale, Mass., assignor to Raytheon Mfg. Co., Newton, Mass., a corporation of Delaware.

This welding system includes a condenser for storing electrical energy, a source of direct current for charging the condenser and a welding load circuit. A controlled ignition discharge tube has a cathode and two anodes with the anodes being connected across the condenser. The cathode in one of the anodes is connected across the input to the welding load circuit and means are provided for igniting the tube.

2,508,115—Impulse Resistance Welder—Hans Klemperer, Belmont, Mass., assignor to Raytheon Mfg. Co., Newton, Mass., a corporation of Delaware.

This welder includes a condenser, a resistance welding load, and means for discharging the condenser into the welding load by an oscillating discharge. The discharge is limited to substantially a single oscillation, and means are provided for operating the discharge means a plurality of times for producing a single welding operation.

2,508,140—Welding Generator—John H. Blankenbuehler, Murrysville, Pa., assignor to Westinghouse Electric Corp., East Pittsburgh, Pa., a corporation of Pennsylvania.

This patent covers a special type of a welding generator.

2,508,145—Welder—Lewis H. Daniels and Henry L. Lindstrom, Detroit, Mich., assignors to Westinghouse Electric Corp., East Pittsburgh, Pa., a corporation of Pennsylvania.

This welder has a unitary timer for timing the current flow throughout a plurality of resistance welding units and a unitary heat control is provided with connections for causing the timer to permit flow of current to each of the units in succession. Additional connections are provided for each timer to preset the heat control at which the timer permits current to flow.

2,508,171—Electrode Construction—Louis W Kaufman, Mansfield, Ohio, assignor to Westinghouse Electric Corp., East Pittsburgh, Pa., a corporation of Pennsylvania.

This patent relates to a combination of supporting plug with an elongated resilient core carried by the plug and having a wire helically wound on the exterior of the core. A plurality of anodic members are spaced longitudinally on the core and have generally cylindrical openings extending therethrough for receiving the resilient core and the wire wound thereon. The resilient core and helical wire have a diameter slightly greater than the diameter of the openings and when longitudinally stretched, a diameter slightly less than the diameter of the openings, so that the anodic member may be moved longitudinally on the core when same is stretched.

2,508,329—Method and Apparatus for Welding Quench-Hardenable Steels—Herbert D. Van Sciver II, Merion, and John J. MacKinney, Narberth, Pa., assignors to The Budd Co., Philadelphia, Pa., a corporation of Pennsylvania.

This welding method relates to welding quench-hardenable carbon-alloy steels and includes the steps of supplying electric current to the weld point in such quantity as normally to produce fusion. Thereafter immediately and without current cut-off the heating current is reduced by instruments from the value at fusion to a value at which the weld temperature is maintained above room temperature and finally the current is reduced to zero.

WELDING RESEARCH COUNCIL
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Supplement to The Welding Journal, July, 1950

Metal Spraying of High-Temperature Metals and Alloys

Technique developed for measuring the strength of sprayed deposits. Relationship exists between strength and oxide content

by Robert T. Thurston and John Wulff

Abstract

The results of an investigation of the metal spraying of high-temperature metals and alloys are presented in this paper. A hydrostatic bursting technique was developed as a means of measuring the strength of sprayed deposits. With this technique, a relationship was shown to exist between the oxide content and strength of sprayed metal.

Sintering of sprayed metal deposits in hydrogen, especially in the presence of a liquid phase, was successfully accomplished, resulting in significant increases in density and bursting strength.

INTRODUCTION

THE current demand for refractory metals and high-melting-point alloys has produced considerable interest in the methods of fabricating such materials. Although attention is presently being focused on techniques such as powder metallurgy, precision casting and die forging, a great need still exists for an inexpensive and simple means of producing large and complex shapes that are difficult to fabricate by virtue of their high melting point, poor machinability or poor hot- and cold-working properties.

The metal-spraying process, heretofore used mainly for coating purposes, merits consideration as a forming technique for large objects for which the powder metallurgical dies and pressures would be economically unfeasible. The spraying technique may be considered a powder metallurgy process except that plastic metal particles are formed into a shape by force of impact rather than ordinary mechanical compaction. Most metals, except perhaps tungsten alloys, may be sprayed by the available techniques.

Although porous metal-spray coats of high-chromium steels and chromium-nickel alloys have been prepared commercially, little work has been reported on either the metal spraying of high-melting-point materials or the production of solid bodies from sprayed metal. Heretofore, metal spraying has been restricted largely to the application of corrosion-resistant coatings and restoration of worn machine components such as shafts and spindles. Ballard[1] refers briefly to the fact that molybdenum and stellite may be sprayed but does not present any data. Likewise, little evidence exists in the literature of the improvement of the poor mechanical properties of sprayed metal deposits by subsequent heat treatment. Although the practice of heat-treating coatings of sprayed aluminum on steel (aluminizing) for purposes of oxidation resistance has been used commercially, Ballard has discouraged the heat treatment of sprayed deposits in general because of volume changes which bring about failure of the coatings by cracking.

This paper describes the spraying of high-melting-point metals and alloys, the properties of the deposits, and the densification achieved by sintering. The

Robert T. Thurston is Research Assistant and **John Wulff** is Professor, Dept. of Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

Table 1—Chemical Composition Before and After Spraying

Metal	*Condition*	*Analysis*						
		Mo	*Ni*	*Cr*	*Fe*	*Co*	*C*	O_2
Molybdenum	Before spraying	99.96	..	..	..	..	0.026	0.009
	After spraying	96.43	..	..	..	..	0.016	3.55
Nickel	Before spraying	..	99.94	..	..	..	0.05	0.008
	After spraying	..	99.72	..	..	..	0.05	0.25
Vitallium	Before spraying	5.79	2.47	27.0	1.48	59.96	0.27	0.058
	After spraying	5.58	2.64	26.3	1.57	59.58	0.19	0.42

materials reported include molybdenum and nickel, molybdenum-nickel alloys and three commercial high-temperature alloys, Hastelloy C, Vitallium and CM-469 (60% Cr, 25% Mo and 15% Fe).

EXPERIMENTAL WORK

Although metal may be sprayed by several different processes,[2] this investigation was confined to the use of the wire-type spray gun for several reasons. The amount of oxidation encountered during spraying is considerably less by this method than by others; a more dense deposit is obtained; and the object being sprayed may be maintained at a considerably lower temperature. Drawn wire or cast rod of the metal or alloy to be sprayed is fed through the center of an oxyacetylene flame by a gear train which is driven by an air turbine. The metal, upon melting, is atomized by a compressed air blast which surrounds the oxyacetylene flame. The sprayed metal particles are still at their melting point when they reach the object being metallized.[3] Upon impact the particles readily collapse, assuming the contour of the surface beneath. Since some oxidation is thus unavoidable, a true sinter bond is not achieved.

Ballard and Harris[4] have measured the percentage of oxygen in sprayed copper produced by four different types of spraying equipment. In their investigation, they observed that the oxide content was a function of the distance between the nozzle of the spray gun and the deposit. It has also been established by Sillifant[5] that the oxide content is affected by the amount of oxygen fed into the flame. Sillifant has shown that if compressed nitrogen is used in place of air, the oxide content is reduced. An inert gas so used, however, is very quickly diluted by oxygen from the surrounding atmosphere.

As shown in Fig. 1, a deposit of sprayed molybdenum is composed of layers of molybdenum metal which are separated by voids and oxide. This is typical of sprayed metals in that a lamellar structure is formed in which oxide is entrapped between layers of metal. That most metals suffer less oxidation than molybdenum during spraying is illustrated by Table 1 which gives the chemical analysis of three of the metals used in this investigation before and after spraying.

With the exception of the sprayed molybdenum samples, the per cent oxygen in each of the above metals was determined by a modification of the vacuum-fusion technique described by Vacher and Jordan[6]

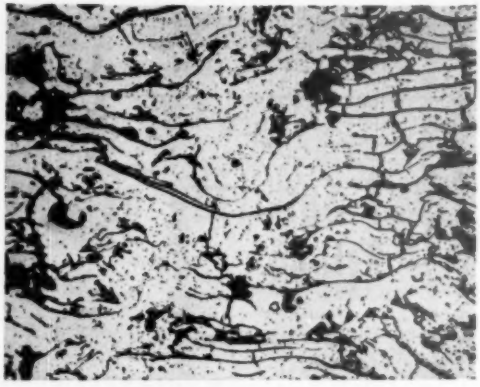

Fig. 1 Cross section of a deposit of sprayed molybdenum; 83% of bulk density. Unetched 200 ×

and Chipman and Fontana.[7] The specimens were melted under vacuum in a graphite crucible containing a bath of iron saturated with carbon. A sample of the gases evolved was analyzed by reacting the hydrogen and CO with CuO to form water vapor, which was absorbed in P_2O_5, and CO_2, which was absorbed in ascarite. The remainder of the gas was considered to be nitrogen. This method of analysis failed to give accurate results in the case of sprayed molybdenum because of the loss of oxygen through volatilization of molybdic oxide prior to melting in the iron bath. Because the density of sprayed molybdenum was so close to that of the iron bath, the molybdenum remained on the top of the bath a sufficient length of time to permit volatilization of some of the oxide. Therefore, the oxygen content of sprayed molybdenum was measured by the weight loss in hydrogen. This technique, which is described in more detail later, gave satisfactory results.

To determine the effect of the oxide content of sprayed metal upon its strength, a method of tensile testing was first developed. Different workers have used various techniques for test specimens. Ballard[1] and Turner prepared tensile test bars by spraying into a steel mold until a deposit was obtained of sufficient thickness to permit machining of test bars. Tubular specimens have been made by Wakefield[8] by spraying a coating of metal on a rod which is later pressed out. In this investigation, production of tubular specimens was first attempted by spraying steel tubes which were later removed by drilling and reaming. This gave considerable trouble because the specimen

cracked in the grips of the tensile testing machine. Hydrostatic bursting was finally adopted in favor of other known types of testing. This has the advantage that test specimens are easily prepared, less metal need be used, the equipment required is relatively simple and the test procedure is readily carried out. The test specimens were prepared on a graphite tube mandrel fitted over a steel shaft $^3/_4$ in. in diameter. The National Carbon Company AGX type of graphite has excellent machinability and gives a smooth finish. After the graphite sleeve was machined to 1.125 in. in diameter, the assembly was rotated in a lathe and sprayed with a coating of metal using $^1/_8$-in. metal wire or rod and a Metco-type spray gun. Throughout the investigation acetylene was used in combination with oxygen and compressed air.

After spraying, the coating of metal was dry-ground to 1.155 in. using a high-speed tool post grinder. A cylinder 2 in. long was then cut from the center portion of the spray coat and the graphite sleeve pressed off the steel shaft. The graphite was removed from the inside of the 2-in. section of sprayed metal and the internal surface paper polished.

This specimen was inserted into the apparatus illustrated in Fig. 2 and the bursting strength determined. The bursting apparatus consists essentially of a thin-walled neoprene tube capped with brass fittings as shown in Fig. 3. Nitrogen gas serves to transmit pressure to the water used for bursting. The pressure is measured with Ashcroft gages. The use of a 2-stage regulator permits measurement over a broad range of pressures.

From the measurements of bursting strength it is possible to calculate the tensile strength of the sprayed deposit. According to Timoshenko,[9] the bursting strength or circumferential tensile stress is given by $\sigma = pd/2t$ where p is the applied pressure, d is the average diameter and t is the wall thickness.

All the as-sprayed samples exhibited a brittle fracture in the hydrostatic bursting test. The results for as-sprayed molybdenum are shown in Fig. 4. The oxide

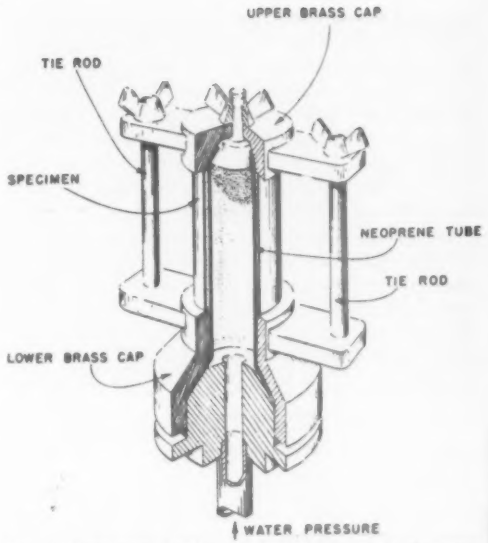

Fig. 3 Specimen holder used in hydrostatic bursting apparatus

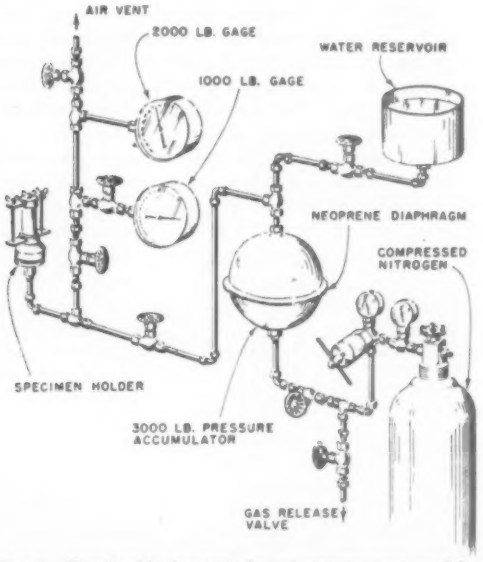

Fig. 2 Sketch of hydrostatic bursting apparatus used for testing sprayed deposits

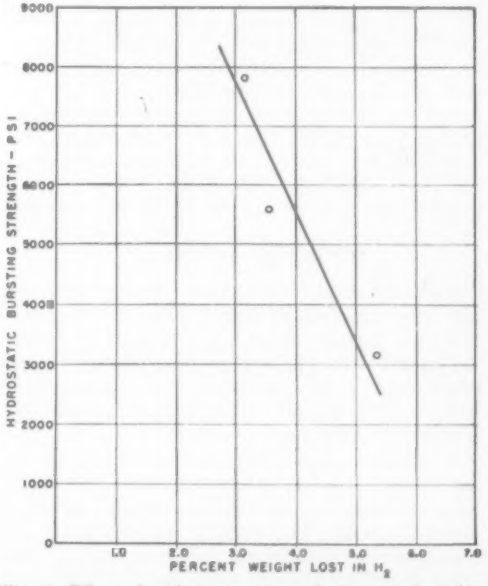

Fig. 4 Effect of oxide content upon hydrostatic bursting strength of sprayed molybdenum

content or weight loss in purified hydrogen at 1300° C. is plotted as abscissa. A small specimen of the fractured cylinder was used for the last-mentioned test, and it was kept at temperature for about 6 hr. or until no further loss in weight occurred. From the curve in Fig. 4 it is evident that the strength of as-sprayed molybdenum is a function of the oxide content. The density of all molybdenum specimens tested was about 85% of that of solid molybdenum.

The fact that this trend is not peculiar to molybdenum is established by similar experiments on sprayed nickel. Here, considerably less oxidation was encountered during spraying, resulting in a deposit of higher density. Again, it is evident that the stronger sprayed product is the one of lower oxide content.

Table 2

Material	*Oxygen %*	*Bursting strength, psi.*	*Density, % theo.*
Nickel	0.46	9,400	91.8
	0.44	10,600	92.7
	0.32	11,100	93.6

Many different alloys were sprayed successfully during the course of this investigation. Among them were Vitallium, which was sprayed from hot-swaged rod, and Hastelloy C and CM-469, which were sprayed from as-cast rod prepared in the laboratory. Melts of the last two alloys were made in a 5-lb. high-frequency induction furnace. The molten metal was drawn from the crucible by vacuum into Vycor glass tubes, producing rods of approximately 12 in. in length. After solidification, the glass was removed from the metal and the short lengths welded together with atomic hydrogen into lengths of 5 ft. or more. In Table 3 the bursting strengths of deposits of these alloys are listed.

Table 3

Metal	*Bursting strength, psi.*
Vitallium	20,000
Hastelloy C	20,100
CM-469 (60% Cr, 25% Mo, 15% Fe)	9,200

SINTERED SPECIMENS

Since the strength of sprayed metals is largely affected by the amount of oxide and the number of voids that they contain, sintering in hydrogen at elevated temperatures was carried out to improve the strength of sprayed deposits. All the specimens were sintered in hydrogen in a $1^1/_2$-in. combustion tube heated in a 2-kw. Globar-type furnace.

To assure a supply of hydrogen free of oxygen and water vapor, the gas was fed through a purification train consisting of tubes of calcium chloride, palladium-alundum catalyst, more calcium chloride, and then a tube of magnesium turnings heated to 400° F. As a rough check on the quality of the hydrogen being used, a ceramic boat containing stainless steel powder was placed inside the combustion tube ahead of the specimen during each run. This aided in indicating the purity of the gas, as well as absorbing any last traces of oxygen.

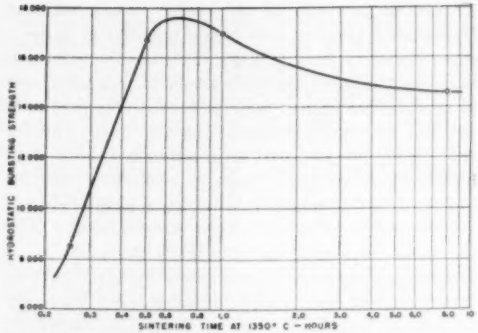

Fig. 5 Variation of hydrostatic bursting strength of sprayed molybdenum with sintering time

To prevent the cylindrical strength specimens from collapsing at the high sintering temperatures, each specimen was fitted with a high-temperature ceramic cylinder which conformed to the internal dimensions of the specimen. This cylinder was readily removed after sintering.

The first molybdenum specimen was sintered for 15 min. at 1350° C. This treatment resulted in only a slight improvement in strength over that of the sprayed deposit. As the sintering time was increased by 15-min. steps, the strength was found to increase. A sintering time of approximately 45 min. at 1350° C. produced the optimum in hydrostatic bursting strength, approximately 17,500 psi. Additional sintering time resulted in decreased strength, as shown in Fig. 5, even though the density of the deposit continued to increase (see Fig. 6). As can be seen in Fig. 7, major reduction of the oxide and elimination of pores were accomplished in the first 15 minutes. The deposits of molybdenum, as well as other sprayed materials, were sintered with no evidence of cracking or distortion.

When nickel specimens were sintered under the above conditions, it was found that a threefold increase in strength over that of as-sprayed nickel could be achieved after only 1 hr. at 1350° C. Although no

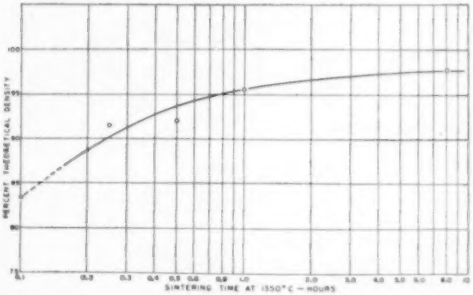

Fig. 6 Variation of density of sprayed molybdenum with sintering time

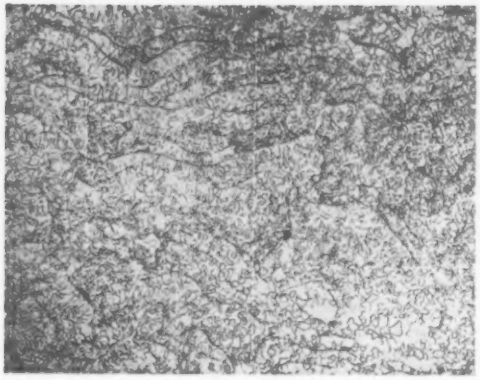

Fig. 7 Sprayed molybdenum after sintering for 15 min. in hydrogen at 1350° C. 250 ×

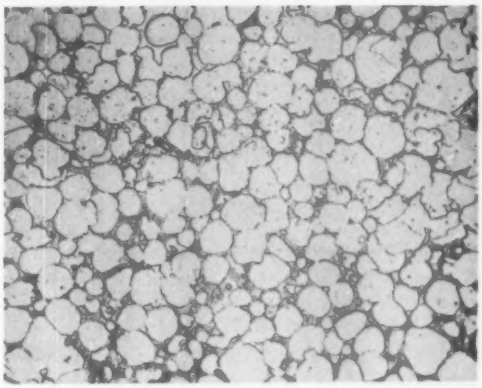

Fig. 8 Sprayed molybdenum with 6% nickel after sintering for 1 hr. in hydrogen at 1375° C. 250 ×

measurements were made of elongation, subsequent examination of the fractured specimen showed considerable increase in ductility after sintering.

Strength measurements were also obtained for cold-drawn nickel and spun molybdenum tubing following a 1-hr. hydrogen anneal at 1350° C. The solid nickel tubes yielded an average bursting strength of 36,000 psi., only slightly greater than that obtained for sintered deposits as shown in Table 4. The tubes spun from molybdenum, however, failed hydrostatically at stresses greater than 54,000 psi. The effect of residual oxygen on the bursting strength of sprayed molybdenum is evidenced by these results.

Even though sintering of sprayed molybdenum resulted in improved properties, sintering in the presence of a liquid phase gave a further improvement. To achieve a liquid phase during sintering of sprayed molybdenum deposits, small amounts of nickel were added by several different techniques. A coating of nickel was first sprayed on the top of a coating of molybdenum. Although some liquid phase was then formed when the composite was sintered, its passage into the molybdenum deposit was limited. A sandwich layer of nickel was found to be impractical because of the void space left by the melting nickel which was too large to be closed by subsequent sintering. The most successful technique was to paint or spray a deposit of molybdenum with successive coatings of a solution of very fine nickel metal powder or nickel salts suspended in alcohol. The porous deposit absorbed the suspension and distributed the nickel throughout the spray coat.

The maximum solid solubility of nickel in molybdenum is only about 0.9% nickel. Above this value an intermetallic compound of extreme brittleness is formed. Consequently, the amount of nickel added to the first two specimens was held below the 0.9% mark. These specimens were sintered for 1 hr. in hydrogen at 1375° C., the lowest temperature at which a liquid phase can be formed. To assure complete reduction as well as densification, the specimens were allowed to remain at temperature for 1 hr. Although additions of 0.25 and 0.78% nickel resulted in strength values greater than sprayed and sintered molybdenum, theoretical density was not achieved. Although this was puzzling at first, metallographic examination showed that such small amounts of nickel were insufficient to completely fill all the interstices in the sprayed deposit. This fact was substantiated when, upon adding 6% nickel, theoretical density was obtained. The photomicrograph of this alloy, presented in Fig. 8, is typical of most liquid-phase sintered alloys. Rounded grains of the solid constituent are surrounded by a matrix of the initially liquefied phase, in this case the "delta" phase. In spite of its high density, the last specimen exhibited very poor hydrostatic bursting strength because of the presence of the brittle delta phase.

The alloys, Vitallium and Hastelloy C, were sintered at a temperature below the liquidus and above the solidus to permit formation of a liquid phase. Figures 9 and 10 illustrate the type of structure which was obtained when these two alloys were thus sintered. It is evident that extensive decarburization took place during hydrogen sintering in both alloys.

Table 4

Metal	*Sintering treatment, 1350° C.*	*Bursting strength, psi.*	*Density, % theo.*
Nickel	15 min.	23,800	96.1
	30 min.	24,150	97.1
	60 min.	30,000	97.6
	480 min.	34,000	98.0

Table 5

Composition	*Sintering treatment, 1375° C.*	*Bursting strength, psi.*	*Density, % theo.*
0.25% Ni	1 hr.	13,250	93.8
0.78% Ni	1 hr.	21,000	95.1
6.0 % Ni	1 hr.	8,000	100.0

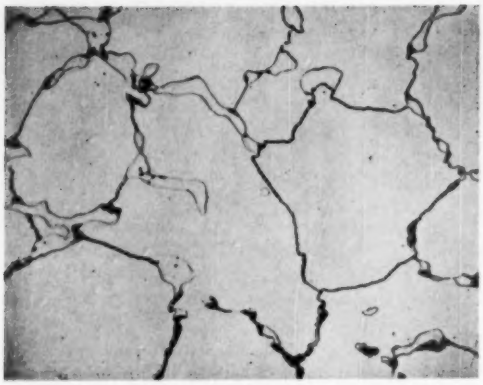

Fig. 9 Sprayed vitallium after sintering for 1 hr. in hydrogen at 1325° C. 250 ×

Because of the extremely high solidus of the CM-469 alloy, somewhat above 1600° C., it was impossible with the equipment available to sinter this alloy in the presence of a liquid phase. Consequently, the strength specimen was sintered at a lower temperature.

Supplementary experiments indicate that densification of such an alloy can possibly be achieved through addition of materials of lower melting point without seriously affecting the properties of the alloy. Additions of small amounts of sprayed iron, for example, permit formation of a liquid phase at a temperature considerably below the solidus of the alloy. As diffusion of the liquid iron takes place, alloying eventually occurs, and the liquid phase disappears. Before this happens, however, appreciable densification is obtained.

Table 6

Metal	*Sintering treatment*	*Bursting strength, psi.*	*Density, % theo.*
Vitallium	1 hr. at 1325° C.	42,000	100
Hastelloy C	1 hr. at 1300° C.	44,000	100
CM-469	1 hr. at 1350° C.	14,030	92.4

Fig. 10 Sprayed hastelloy C after sintering for 1 hr. in hydrogen at 1325° C. 250 ×

DISCUSSION OF RESULTS

The method of forming a thin-walled body of rotational symmetry by metal-spraying followed by sintering in a reducing atmosphere permits the fabrication of bodies impossible or difficult to form by powder metallurgy, casting or forging techniques. To achieve optimum mechanical properties, maximum density is imperative. The porosity and the oxide content of the as-sprayed body limit its usefulness. Densification and removal of oxide are only feasible if a liquid metallic phase of sufficient quantity is formed during sintering. The present research indicates that with sprayed deposits of pure metals it is difficult, if not impossible, to achieve bulk density. Furthermore, the total removal of oxide by hydrogen reduction in such cases is hindered by the difficult removal of entrapped water vapor in the later stages of sintering.

The experiments described in this paper clearly illustrate that the presence of sufficient intergranular liquid phase during hydrogen sintering serves to eliminate voids and oxides. With pure metals this usually demands the addition of another metal which forms by alloying a low-melting-point phase at a sintering temperature below the melting point of the parent metal. With sprayed alloy coatings the temperature of sintering must be adjusted between the solidus and the liquidus.

Certain shapes of nonrotational symmetry can also be sprayed and sintered. In sintering such shapes, shrinkage cracks readily develop unless sufficient liquid phase is formed before appreciable densification occurs. On the other hand, the stresses developed during the shrinkage of deposits on flat surfaces are sufficient to either peel or crack the deposit. This may even occur during spraying of deposits of appreciable thickness on plane surfaces.

Densification of sprayed metal deposits by subsequent heating with oxyacetylene or atomic-hydrogen arc flames was found possible but in most cases not practical. Oxide removal proved more difficult, and cracking of the deposit occurred frequently. If a flux was simultaneously sprayed with the metal or if an alloy addition of boron was contained in the spray coat, flame sintering from the standpoint of densification was more successful. Nevertheless, the mechanical properties achieved did not compare with those reported in this paper.

A technique of simultaneously spraying and sintering in air was also investigated with various commercial spray guns, as well as with those of our own development. In these experiments metal powders, as well as mixtures of powder and flux, were used. For hardfacing or coating purposes a fair degree of success was obtained, but in the fabrication of hollow bodies of significant dimensions these methods were only of limited usefulness.

Although only bodies of test size are reported in the present paper, it is possible to employ the technique described for fabrication of bodies of appreciable size. Sintering of the whole body in electrical resistance

furnaces then proves difficult. It has been found more useful, however, to sinter the body in quartz tube furnaces where the work is heated by induction. In one design of this type of furnace, the quartz tube containing the work was held stationary while the high-frequency coil was moved. Here, heating is started at the top of the body and the induction coil moved downward. Thus, a narrow annular region of the deposit is sintered and the coil moved to an adjacent position at a predetermined rate before extensive melting and distortion occur. Here again, the presence of some liquid phase is imperative if maximum density is desired. Induction sintering of this kind is more rapid and economical than resistance furnace sintering. Furthermore, it can be carried out in vacuum, as well as with reducing or inert gases.

In the previous section results for molybdenum-nickel-alloy spray deposits were used to illustrate the advantage of liquid-forming additives. Unfortunately, for many applications the intergranular delta phase is too brittle if 6% nickel is employed. Although complete densification may be achieved with smaller amounts of nickel, the intergranular delta network still remains somewhat brittle. Cobalt and iron, which likewise form low-melting phases, appear from auxiliary work to be less embrittling. Molybdenum deposits of this kind can be further strengthened by using a molybdenum-tungsten spray coat made from alloy wire instead of ordinary molybdenum. Tungsten in solid wire form was found impossible to spray with the apparatus available.

From an economic standpoint the use of cast or swaged rod or wire of high-melting-point material is for some purposes prohibitive. Efforts to obtain equivalent quality spray deposits with extruded wire made from metal powder and plastic binder, a far cheaper material, did not prove successful for molybdenum- and tungsten-rich deposits. When such extruded wire is presintered to remove organic binders and to weld the constituent metal powder particles together, a fragile but more useful product is obtained. This technique also permits the inclusion of high-melting-point nonmetallics.

The general process of spraying metals or alloys and subsequent sintering for maximum densification is, like most powder metallurgy processes, only applicable to fabrication of products which either cannot be made otherwise or as cheaply. For this reason the process has but limited application and is best suited to the fabrication of thin-walled articles of rotational symmetry made from high-melting-point materials.

In summary, the experimental work reported in this paper shows that dense hollow bodies of high-melting-point alloys can be fabricated by spraying and sintering in hydrogen if a liquid phase is present during the latter heat treatment.

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Welding Research in The Netherlands

Since Jan. 15, 1948, the Netherlands Welding Society has been cooperating with the Belgian Welding Society in research on the weldability of steel. The first two reports of the research were published in April 1950.

The research program consists of notch-tensile, notch-bend and notch-impact tests (Charpy and Schnadt) of two grades of steel, in an attempt to determine the causes of brittle fractures. Two steels are being used. The first steel is a killed and normalized open-hearth steel 1.57 in. thick with very good weldability and having a tensile strength of 53,000 to 64,000 psi. The second steel is an unkilled basic Bessemer steel 0.79 in. thick with average weldability and having a tensile strength of 53,000 to 64,000 psi. Owing to delay in securing the two steels, the Netherlands Committee performed its preliminary work on a steel obtained from the Royal Netherlands Blast Furnace and Steel Co.

In the preliminary work an apparatus was built to maintain notch impact specimens at low temperatures in a mixture of alcohol and dry ice. Experiments were made to determine the change in temperature of impact specimens, cooled below room temperature, during the period between the moment of removal from the cooling bath and the moment of being hit by the pendulum. It was found that for bars cooled to −60° C. the change in temperature is negligible if the period in question does not exceed six seconds.

The Research Committee of the Netherlands Welding Society is composed of Dr. H. G. Geerlings, *Chairman*, Dr. J. A. Haringx, *Secretary*, Prof. W. F. Brandsma, J. Palm, and R. Reitsema. The experimental work is conducted in the Metals Division of the Central Institute for Materials Research.

Determination of the Temperature of Sprayed Metal Particles

Heat content and temperature of sprayed iron and nonferrous particles at the moment of impact

by James E. Cline, Robert T. Thurston and John Wulff

Abstract

This paper presents the development of an experimental technique for determining the heat content and temperature of sprayed metal particles at the moment of impact. The investigation includes tests made with zinc, aluminum, copper, nickel, iron and molybdenum.

The results obtained indicate that the metal particles are at their melting point at impingement. Calculations are included to show that the contribution of the kinetic energy to the heat content and temperature of the particles is a negligible quantity.

INTRODUCTION

THE first attempt to explain the physical character of the metallic particles produced by the metal-spraying process was presented by Schoop.[1] He believed that a large part of the kinetic energy imparted to the metallic particles by the surrounding air blast was converted to heat as the particles impinged upon the material being sprayed. According to Schoop, the particles were solid during transport, having cooled very rapidly after leaving the gun, and became plastic momentarily upon striking. In this condition they were welded into a firm layer.

A few years after Schoop published his explanation, Arnold[2] calculated the speed necessary to give sufficient energy to a particle to permit transformation from the solid to the plastic state and compared this speed with actual measured speeds of sprayed particles. He disproved Schoop's theory by showing that the actual speed of sprayed particles was far below that required to produce a temperature rise of the necessary magnitude. As further proof for his theory Schoop pointed out that the hand, held a few inches from the nozzle of the spray gun, experienced only a feeling of warm air blast and that combustible materials could be metal-sprayed without danger from the heat of the particles. It has been shown since then that in the case of high-melting-point materials this is true only when spraying from a great distance.

That the temperature of the base material actually increases during spraying was shown by Thormann,[3] who measured the temperature of a metal sheet with a thermocouple fastened to the back of it while it was being sprayed with aluminum and iron. A rapid increase in temperature was detected at the beginning of spraying; but as the thickness of the deposit increased, the temperature rise became less until a constant temperature was obtained, varying from 130° C. for aluminum to 220° C. for iron.

Another explanation is that of Karg,[4] Kutscher[5] and Reininger.[6] They assume that the cold, solid particles are hammered into the pores of the surface by the force of the air blast behind them. Karg calculated the temperature of air undergoing adiabatic expansion to show that the particles are cooled by the cold expanding air. Although he arrived at a temperature of 63° C. from an original pressure of 2.2 atmospheres and an original temperature of 20° C., his calculations were later shown to be in error because of failure to consider that appreciable pressure loss occurs in the gun itself and that the air immediately after expansion passes through the hot portion of the flame where it undergoes considerable heating. Experiments by Thormann[3] clearly refute both Schoop's and Karg's theories. By use of an optical pyrometer in a darkened room he measured the temperature of sprayed iron particles and found them to be considerably above the temperature predicted by Karg. At a distance of 10 to 20 cm., values around 1000° C. were obtained.

A third explanation, presented by Schenk,[7] assumes that the temperature of an impinging particle lies above its melting point. This theory is difficult to comprehend on the basis of known facts. Since the particle leaves the gun at a very rapid rate after it melts, it is difficult to conceive of any degree of superheating occurring in that short space of time. Again, the work of Thormann seems to contradict this theory. Thormann claims that the origin of sprayed metal deposits may be explained as follows: The metal, after melting, is only slightly cooled by the air blast, since the air itself is strongly heated by the flame and the metal remains in the air stream an extremely short

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time. Consequently, the particles are very hot, weak and plastic when they strike but are not above the melting point. Turner[8] supports Thormann's explanation on the basis that the particles are moving at the same speed as the air blast and therefore suffer little cooling effect from it. He has calculated for zinc that a time of 0.0014 sec. is required for a particle to travel a distance of 20 cm. from the nozzle of the gun.

Although the general consensus of opinion among investigators is that the particle must be plastic to permit formation of the typical laminated sprayed structure, there is no agreement as to the temperature of the particle at the moment of impact. The purpose of the present investigation was the experimental determination of this temperature.

EXPERIMENTAL WORK

In this investigation the mean temperature of the metallic particles at the time of impingement on the base material was calculated from the heat that was carried to the surface by a known weight of the metal. The principle employed in the experimental measurement was that of balancing the heat input from the particles with the heat input from electrical energy to produce the same equilibrium temperature under the same air-flow conditions. As this technique depends upon a calorimetric determination of the heat content of the particles and not upon a direct measurement of their temperature, many of the influencing factors which confronted former investigators have been eliminated.

As shown in Fig. 1, the apparatus consisted of a 2-in. square of ceramic plate $^1/_4$ in. thick on which a heating coil of 200-w. capacity was mounted. The coil was fixed to the ceramic plate with alundum cement and covered with a thin sheet of mica to insulate it from the $^1/_8$-in. copper plate which was used as the spraying surface. Four small clamps held the assembly together. A chromel-alumel thermocouple was inserted through a hole in the center of the ceramic plate and firmly imbedded in the copper plate. The sides and back of the assembly were covered with asbestos insulation to prevent undue heat loss. A recording Esterline-Angus d.-c. milliammeter gave a permanent record of the thermocouple output; a Weston ammeter and voltmeter indicated the power input to the heating coil.

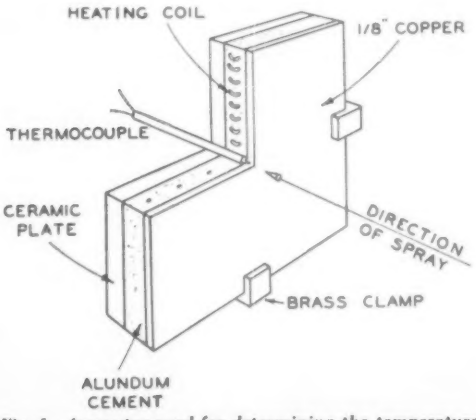


Fig. 1 Apparatus used for determining the temperature of sprayed metal particles

The air-driven, wire-type spray gun which was used throughout the investigation was mounted on a support which enabled the operator to maintain a constant spraying distance from one run to the next. When the gun was set in operation and final adjustments completed, it was moved in front of the test apparatus and allowed to deposit metal on the surface of the copper plate until a constant temperature was obtained, as indicated by the recording milliammeter. This temperature shall be referred to as the *saturation temperature*. At this point, the heat supplied to the plate by the depositing metal is equal to the heat lost from the copper plate through conduction and radiation. It is assumed that the temperature gradient between the surface of the sprayed deposit and copper plate is negligible.

As the success of this type of measurement is dependent upon the efficiency with which heat is transferred from the deposit to the copper plate, it was necessary before each run to sandblast the surface of the copper plate thoroughly to assure a sufficiently rough surface to provide good adhesion of the deposit to the plate. Unless considerable care is taken in this step, the corners or outer edge of the deposit tend to pull away from the copper plate under the action of the stresses which are induced upon cooling. At the saturation temperature, heat is supplied to the deposit from two sources—from the gases of combustion which impinge on the deposit and from the metal particles themselves. If the heat supplied by the metal particles can be replaced by another measured heat input to give the same saturation temperature, the heat content of the particles may be determined. This is done by repeating the above process with the wire removed from the spray gun so that only the air blast impinges on the deposit. The heat which was formerly supplied by the depositing metal is now supplied by the heating coil located behind the deposit. The power input is gradually increased until the temperature of the copper plate corresponds to the saturation temperature. The heat contributed by the heating coil is now equivalent to the heat formerly supplied by the metal particles and is a measurable quantity.

After the deposit has cooled to room temperature, it is removed from the copper plate and weighed. The spraying time is obtained from the strip chart on the Esterline-Angus milliammeter, and the rate of deposition is calculated from these two figures. The calories per mole of deposited metal are determined from the rate of deposition and from the amount of heat supplied by the heating coil. Using specific heat data by K. K. Kelley, Bureau of Mines Bulletin No. 476, the average

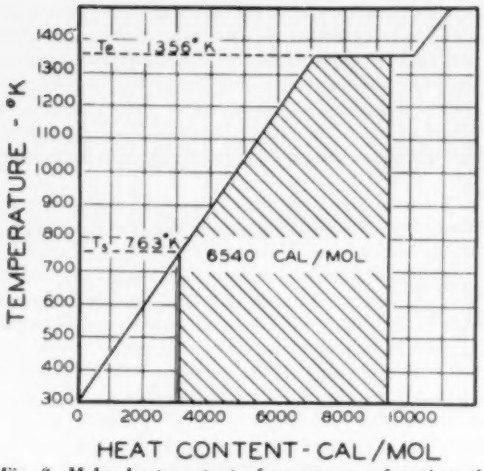


Fig. 2 Molar heat content of copper as a function of temperature

temperature of the sprayed particles is calculated as shown below.

Copper

$$\text{Rate of deposition} = \frac{\text{weight}}{\text{time} \times \text{mol. wt.}} = \frac{67.1}{183.6 \times 63.57} = 0.00575 \text{ mol./sec.}$$

Heat input = watts × 0.239 = 37.64 cal./sec.

$$\text{Heat content} = \frac{37.64 \text{ cal./sec.}}{0.00575 \text{ mol./sec.}} = 6540 \text{ cal./mol.}$$

T_s = saturation temp.

T_p = particle temp.

$$\Delta H = \int_{T_s}^{T_p} C_p dT$$

$$\Delta H = \int_{T_s}^{T_p} (5.41 + 1.50 \times 10^{-3} T) dT$$

$$= 5.41 (T_p - 763) + 0.75 \times 10^{-3} (T_p^2 - 763^2)$$

At T_p = 1356° K.

ΔH = 4150 cal./mol.

Heat of solidification, ΔH_s = 3120 cal./mol.

Since 4150 + 3120 > 6540 cal./mol., not all of the copper can be molten, and therefore T_p = 1356° K. or 1083° C.

These calculations show that part of the heat content given up by each particle comes from the heat of solidification, as illustrated in Fig. 2.

Of the six pure metals tested, all were sprayed at a distance of 5½ in. from the test apparatus with the exception of iron and zinc which were sprayed at 8 in. In the case of iron it was found that if the spray gun was held only 5½ in. from the copper plate, the temperature of the deposit was sufficient to permit the occurrence of excessive oxidation which liberated enough heat to cause localized melting of the deposit. By increasing the spraying distance to 8 in., this condition was alleviated. In the case of zinc, the lowest melting metal tested, the temperature of the impinging flame itself was high enough at 5½ in. to cause the deposit to melt and drip from the copper plate.

The experimental results and calculated temperatures of sprayed particles of copper, molybdenum, aluminum, iron, nickel and zinc are listed in Table 1.

DISCUSSION

The initial runs which were made on copper were consistent in indicating the melting point as the temperature of the sprayed particles at the moment of impact. Results of the tests made on aluminum and zinc were similar in that the particles of these two metals were also found to be at their melting point. When higher-melting-point materials were tested, considerable difficulty was encountered in securing an adequate bond between the copper plate and the deposit. Consequently, some values were obtained below the melting point for nickel, molybdenum and iron. When good adhesion was obtained, however, the particles were found to be at their melting point.

The kinetic energy of the particles may be calculated from the data given in Table 1 if the velocity of the particle is known. Arnold[2] measured the velocity of brass particles at a distance of 10 cm. from the nozzle and found the mean value to be approximately 120 meters per second. It is assumed that this figure holds for copper. During Run No. 1, copper was deposited at a rate of 0.366 gm./sec. At this rate the energy of impact was 2.64×10^7 ergs/sec. or 0.630 cal./sec. The total heat contributed by the particles, as measured in Run No. 1, was 37.7 cal./sec. Thus, the energy of impact is equivalent to only 1.7% of the total heat.

In the preceding calculations, a total of 6540 calories was contributed by each mole of copper. Of this amount, 4150 calories were obtained as "sensible heat"

(*Continued on page 342-s*)

Table 1—Temperature of Sprayed Metallic Particles Determined Calorimetically

Run	*Metal*	*Melting point, ° C.*	*Saturation temperature, ° C.*	*Power, watts*	*Dep. rate, gm./sec.*	*H, cal./mol.*	*Tp, ° C.*
1	Copper	1083	490	157.5	0.366	6,540	1083
2	Copper	1083	504	157.5	0.386	6,200	1083
3	Molybdenum	2625	452	187.0	0.252	17,020	2625
4	Molybdenum	2625	490	185.0	0.297	14,300	2390
5	Aluminum	659	476	82.5	0.216	2,460	659
6	Aluminum	659	476	106.0	0.243	2,810	659
7	Iron	1530	500	180.0	0.177	13,600	1530
8	Iron	1530	469	135.0	0.195	9,240	1370
9	Nickel	1452	539	190.0	0.305	8,740	1452
10	Nickel	1452	493	93.0	0.231	5,650	1200
11	Zinc	420	375	86.4	0.863	1,560	420
12	Zinc	420	377	84.0	0.651	2,020	420

Shrinkage Distortion in Welding

A Review of the Literature from Jan. 1, 1944 to July 1, 1949 (Foreign Literature from Jan. 1, 1941). This report is prepared under the auspices of the Structural Steel Committee of The Engineering Foundation—Welding Research Council

by W. Spraragen and W. G. Ettinger

SUMMARY

DISTORTION in welding is due to the natural forces of expansion and contraction, which are a function of the particular pattern of heat distribution throughout the piece being welded. Furthermore, concurrent with an increase in temperature, metals suffer a decrease in their yield point, that is in their resistance to plastic deformation. For mild steel, which retains its normal physical properties from room temperature up to about 600° F., the yield point drops to 5000 to 10,000 psi. at 1100° F. Thus, it is not only the weld metal which becomes permanently deformed as a result of the welding heat, but a considerable portion of the base metal near the weld zone as well. Beyond a certain distance from the weld, where the temperature is such that the yield point equals or exceeds the thermal stress, the deformation is of purely elastic, that is of a temporary nature. The most obvious objections to distortion is from the viewpoint of holding dimensions, proper fit and physical appearance, but is equally serious if cracks due to shrinkage are to be avoided. While a considerable reduction in welding distortion can be effected by the adoption of a judicious overall procedure, properly selected types of joints, as well as peening subsequent to welding, shrinkage allowances need be made in the design of most welded structures.

BUTT WELDS

Transverse Shrinkage

Distortion due to shrinkage transverse to the direction of the weld depends principally on the cross sectional area of the weld, or more generally, the amount of weld metal deposited. Thus, the larger the weld cross section, the greater is the shrinkage. This contraction can be estimated from a chart showing transverse shrinkage as a function of the weld cross section for various plate thicknesses, or from an empirical formula which incorporates these variables. Other influencing factors are the degree of applied restraint, length of the joint, number and size of layers, welding position, speed of weld travel and type of process used. There is general agreement that the greater the restraint, the less is the transverse shrinkage. It takes a certain minimum length of weld for transverse shrinkage to become stabilized for a given set of conditions.

Longitudinal Shrinkage

Shrinkage parallel to a butt weld is of special importance in the welding of sheet metal, since it is the chief contributor to buckling along the joint. It is best minimized by confining the heat to a narrow zone and by properly designed jigs. Few data on longitudinal shrinkage in butt welds appear in this Review, but, in general, the summary of longitudinal shrinkage in fillet welds is equally applicable to butt welds.

FILLET WELDS

Transverse Shrinkage

Transverse shrinkage in fillet welds is largely confined to the plane of the plate and increases with the total area of the fillets. While the restraint normally increases with plate thickness, this effect is here generally offset by the use of correspondingly larger fillets. For the most part, transverse shrinkage in the plates is materially less for fillet welds, as compared with shrinkage for a corresponding weld cross section in butt welds. In welding several stiffener plates to the same through plate, the subsequent stiffeners may cause a readjustment in the shrinkage in the through plate due to the prior plates, but after a certain number of stiffeners the total shrinkage tends to become stabilized.

Longitudinal Shrinkage

Shrinkage parallel to a fillet weld is a function of the transverse cross section of the weld and of the transverse cross section of the colder portions of the assembly resisting the expansion and contraction forces of the hot metal during welding and cooling. Thus, shrinkage, which attains high values for small resisting cross sections, decreases rapidly as the section increases and tends to become stabilized when the resisting cross section exceeds a certain value. Longitudinal shrinkage is proportional to the length of the weld.

WELD BEADS

Beads of weld metal deposited on a flat plate resemble fillet welds. Thus, longitudinal shrinkage is in direct proportion to the heat input or, by extension, the amount of weld metal deposited. Deformation due to weld beads laid unsymmetrically with respect to the neutral axis of a member varies with the width (depth) of the member and also the welding method used.

ANGULAR DISTORTION

Angular distortion, that is the lifting of the free ends of two members being joined out of their initial plane, with the weld as the axis of rotation, is a result of unsymmetrical shrinkage forces acting across the width and/or through the thickness of the members at the joint. It occurs in butt as well as fillet welds, and is a function of the number and size of the layers deposited, the interpass temperature and the length of the weld. There appears to be universal agreement that angular distortion increases with the number of layers for a given size of total weld deposit. Angular distortion can be effectively controlled by presetting the members against the anticipated distortion, by a balanced welding sequence, by preventive supports or by proper joint design. In butt welds, U-welds and double V-welds produce less angular distortion than single V-welds.

W. Spraragen and W. G. Ettinger are respectively Director and Technical Assistant Welding Research Council.

SHRINKAGE DISTORTION IN WELDING

Reviews summarizing published information on the subject of shrinkage and shrinkage distortion due to welding up to Jan. 1, 1944 (foreign literature to Jan. 1, 1941) have already appeared. The present report covers the literature which has been published in the intervening period up to July 1, 1949.

The vast majority of investigations reported on shrinkage distortion deals with the welding of ferrous metals, with prominence given to the plain carbon steels. It should be emphasized that this review is not applicable to nonferrous materials for reasons of (1) lack of information on their behavior during welding, and (2) decidedly different thermal properties of these materials. For the most part, the information on distortion due to welding is qualitative rather than quantitative and is concerned mainly with practical methods of distortion control. During the war years and those immediately following the war, a considerable number of articles were written on methods of shrinkage control and recommended welding sequence in ship building. Since papers of that type invariably involve very specialized applications and are of a very detailed nature, it was felt that a review of these would lie outside the scope of this report. For similar reasons, the subject of shrinkage stresses caused by welding was not included in this discussion.

The present Review consists of a summary of the shrinkage distortion for all types of welding, reported in the literature, with first consideration given to an analytical treatment thereof.

BUTT WELDS

Transverse Shrinkage

Distortion due to shrinkage transverse to the direction of welding depends principally on the cross-sectional area of the weld. This was already brought out in two previous Reviews of the Literature.[1, 2] Further evidence of this relationship is furnished by a number of authors who have investigated the shrinkage problem more recently.

Guyot[3] presents a chart, Fig. 1, which permits a quantitative evaluation of transverse shrinkage in unrestrained plates of various thicknesses. Therein experimental shrinkage values are compared with those calculated by means of an empirical formula. This formula was originally developed by Malisius and has already been covered in a previous Literature Review.[1] It is repeated here for the reader's convenience:

$$S = 0.1716\,\frac{A}{t} + 0.0121w$$

where

S is the transverse shrinkage, in.

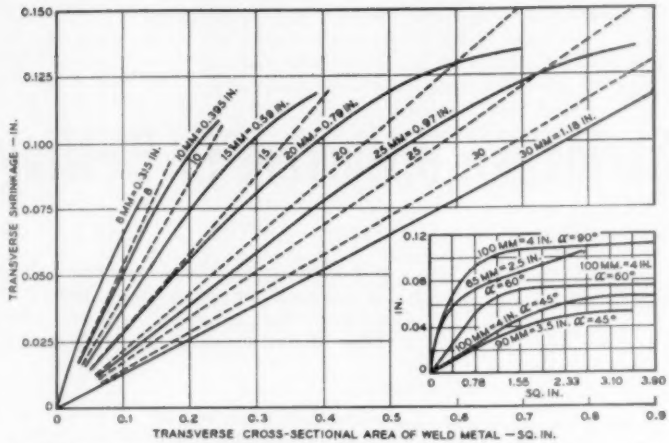


Fig. 1 Transverse shrinkage of butt welds as a function of the cross-sectional area of the weld for different thicknesses of plate. Guyot[3]

A is the cross section of the weld, sq. in.
t is the average plate thickness, in.
w is the average width of the weld, in.

The solid curves in Fig. 1 represent the variation of transverse shrinkage with the transverse cross-sectional area of the welded joint for a series of plate thicknesses. The dotted lines correspond with the shrinkage calculated by means of the empirical formula.

The plates were welded with covered electrodes. Full weaving was used for the thinner plates and a combination of weaving and beading for the thicker plates. The plates were beveled for V or double V and with varying root spacing to vary the sectional area of the weld. The electrodes were 0.128, 0.16, 0.20 or 0.24 in. diameter, depending on the joint, and the welding currents were 150 amp. for 0.128 in., 190 amp. for 0.16 in., 250 amp. for 0.20 in. and 310 amp. for 0.24 in. The assemblies were 11¾ by 11¾ in. Shrinkage was measured between three pairs of lines 4, 8 and 11¾ in. apart. Three points on each line, one at each end and one at the middle, permitted nine measurements on each assembly. The speed of welding was normal, and interruptions were just sufficient to prevent abnormal heating of the plates.

Since the plates were completely free to move (details of support not stated), the measured shrinkages are the maximum to be expected and even exceed the calculated values for plates up to 1 in. thick. According to the author, for plates thicker than 1 in. the internal stresses are such that the shrinkage becomes less and less. At large values of weld cross section the shrinkage tends to become constant and in no case exceeded 0.142 in.

Guyot also includes several practical examples, in which he demonstrates the use of this chart and a similar chart de-

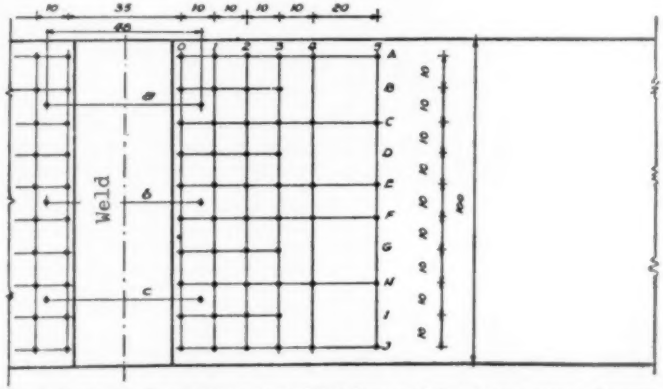


Fig. 2 Specimen for measurement of shrinkage and shrinkage stresses. Campus[4]

Lines a, b and c are gage lengths of 1.9 in. for shrinkage measurements. The solid black circles represent balls pressed into the plate to serve as gage marks for shrinkage stress measurements. Dimensions in millimeters.

veloped for estimating longitudinal shrinkage (see section on longitudinal shrinkage in fillet welds in this Review) as applied to the welding of locomotive subassemblies.

Fairly good agreement with Guyot's results is obtained by comparison with those of Malisius and Berriman, discussed in two previous Literature Reviews,[1, 2] as far as they are applicable.

Further agreement, but limited to a single plate thickness, is found by comparing them with those of Campus[4] who made a comprehensive study of transverse shrinkage as a function of the mean width of the joint in welding 0.55-in. basic Bessemer steel plate. The specimens were 4 in. wide and 12 in. long, the butt joint being transverse to the direction of rolling, Fig. 2. Welding was done with good commercial electrodes. For most tests the root passes were deposited with 0.128-in. electrodes, the remaining passes with 0.160-in. electrodes. Most of the measurements were made with a Zeiss measuring microscope, but Huggenberger tensometers and a Mahr type of mechanical extensometer were also used. The plates for unrestrained tests were tacked 1/2 in. from each end of the joint and welded in the flat position, being turned 180° after each pass to equalize the heat. Double V-joints were welded alternately on each side. A 1/8 x 5/8 in. backing strip was used for the V-joints. The author reports that measurements were the same whether the backing strip was left on or machined away after welding.

Campus' results of the tests on unrestrained butt welds in 0.55-in. plate are plotted in Fig. 3. The shrinkage is approximately proportional to the average width of the joint, for V- and double V-joints with 60, 70, 80 and 90° openings, and root spacings varying from 0.04 to 0.18 in. From this plot Campus derived the formula $P_m = 0.179\ L$, with a probable error of ±14.4%, where P_m is the transverse shrinkage, and L is the average width of the joint, both measured in inches. He found, however, that this empirical formula holds only for V-joints in the range from 0.20 to 1.2 in., and for double V-joints only in the narrow range from 5/8- to 7/8-in. plate thickness. He shows the effect of plate thickness on transverse shrinkage in Fig. 4, but fails to account for the abnormally high shrinkage in double V-joints in the range from 0.40- to 0.71-in. plate thickness, shown to be considerably greater than that in single V-joints.

Campus also investigated the effect of restraint on transverse shrinkage in welding 0.55 in. plate. The specimens were mounted into a heavy frame in such a manner that the effective distance from butt joint to frame was 19 in. The shrinkage was very small before the specimens were cut loose from the frame. After being cut loose, they showed only about 50% of the shrinkage of the unrestrained butt welds.

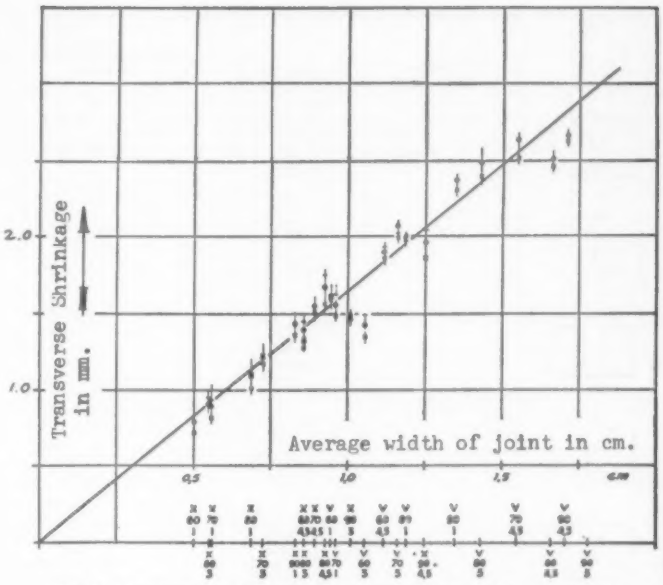


Fig. 3 Transverse shrinkage of unrestrained butt welds. Campus[4]

Vertical scale: Average of shrinkage measured on face and back of joint, millimeters. Horizontal scale: Average width of joint, centimeters. V- and X-joints with 60, 70, 80 and 90° included angles, and 1-, 3- and 4.5-mm. root gap are plotted. ○, measurements made on gage length b, Fig. 2. x, measurements made on gage lengths a and c, Fig. 2

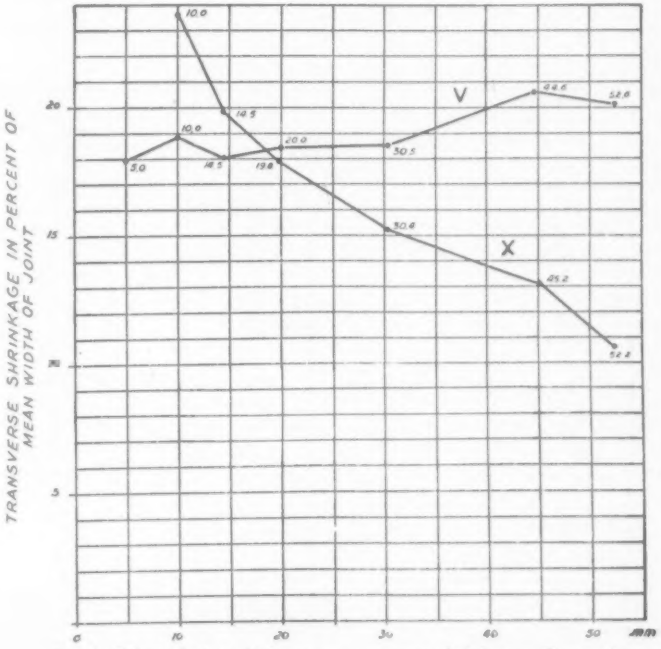


Fig. 4 Effect of plate thickness on transverse shrinkage. Campus[4]

Vertical scale: Transverse shrinkage in per cent of mean width of joint. Horizontal scale: Thickness of plates, millimeters. The joints were 6 in. long 80° V and double V (X) with 1/8-in. root gap.

Weld	Position in welding	Sequence of weld deposits	Measured transverse shrinkage, mm.	Measured transverse shrinkage, in.	Sectional area of weld, mm.²	Sectional area of weld, in.²	Shrinkage calculated according to empirical formula, mm.	Shrinkage calculated according to empirical formula, in.
	Vertical		2.27	0.089	324	0.51	1.8	0.071
	Same		2.97	0.117	425	0.66	2.37	0.093
	Same		3.3	0.130	488	0.76	2.72	0.107
	Same		2.83	0.111	417	0.65	2.33	0.092
	Same		3.25	0.128	493	0.765	2.74	0.108
	Same		3.47	0.137	565	0.87	3.15	0.124

Fig. 5 Effect of weld cross section on transverse shrinkage in 1.3-in. plate. Tsegel'skii and Mel'nik[5]

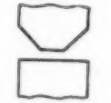
Weld	Position in welding	Sequence of weld deposits	Cross sectional area of weld, mm.²	Cross sectional area of weld, in.²	Measured transverse shrinkage, mm.	Measured transverse shrinkage, in.	Shrinkage calculated according to empirical formula, mm.	Shrinkage calculated according to empirical formula, in.
	Horizontal on vertical plane		423	0.66	2.31	0.091	2.36	0.093
	Same		461	0.71	2.48	0.098	2.57	0.101
	Same		500	0.77	2.76	0.109	2.78	0.109
	Same		499	0.77	2.60	0.102	2.78	0.109
	Same		649	1.0	3.24	0.127	3.72	0.146

Fig. 6 Effect of weld cross section on transverse shrinkage in 1.3 in plate. Tsegel'skii and Mel'nik[5]

The effect of the number of passes on transverse shrinkage in 0.55-in. thick plate, for a 90° V-joint with a 0.12-in. root gap, is shown by Campus in Table 1. The welding was done with 0.160-in. electrodes. It is not stated, however, whether weaves or beads were used.

Table I—Effect of Number of Passes on Transverse Shrinkage of 90° V Butt Joints. Campus[4]

Number of passes	*Transverse shrinkage, in.*
7	0.128
9	0.144
10	0.156
12	0.174
15	0.182

Varying the current from 160 to 205 amp. was observed to have little effect on transverse shrinkage.

Campus found the interpass temperature to have only a negligible effect on transverse shrinkage in butt joints. The length of the joint, however, produced a considerable variation in transverse shrinkage. Thus, for an 80° V-joint with 1/8-in. root gap in 0.55-in. plate, the shrinkage measured 0.098 in. for a joint 4 in. long and 0.108 in. for a joint 6 in. long. For an 80° double V-joint the shrinkage was 0.052 in. for a joint 4 in. long and 0.071 in. for a joint 6 in. long.

Tsegel'skii and Mel'nik[5] conducted a series of experiments on transverse shrinkage in 1.3-in. thick steel plate in preparation for the construction of a welded blast furnace jacket. Different sizes of butt joints and two different welding positions were used. Three measurements, taken before and after the welding across the transverse width of the plates, were made by means of a strain gage consisting of a dial indicator and a ground cone-shaped support mounted on the opposite ends of a long bar. The accuracy obtained was 0.008 to 0.003 in.

The plates, 1.3 in. x 1.3 ft. x 3.3 ft. in size, were butt welded with vertical and horizontal welds along the longer dimension. 45 and 60°, double V-joints were used for the vertical welds, and K-shaped joints with angles of 45 and 55° for the horizontal welds, while the root spacings varied from 0.08 to 0.24 in. The specimens were assembled without clamps, on wedge-shaped positioning devices. The vertical welds were made in three layers from each side by the back step method, whereas the horizontal welds were finished in six to ten layers from each side. Type 6010 electrodes and normal welding conditions (no details) were used.

The results of the shrinkage measurements, obtained with the double V- and K-joints, respectively, and compared with theoretical values calculated according to Malisius' empirical formula, (stated earlier in this Review) are shown in Figs. 5 and 6.

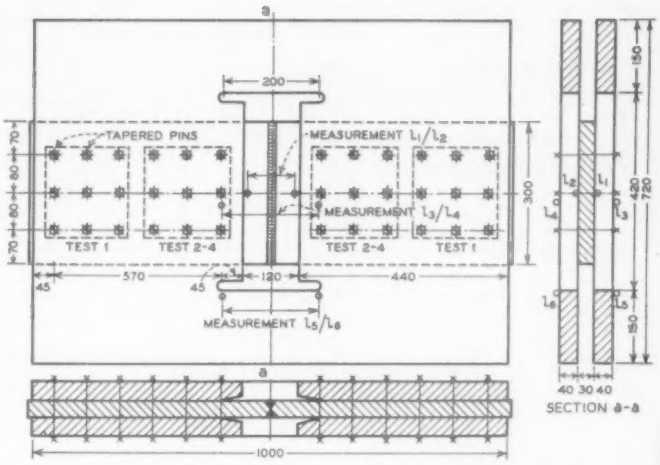


Fig. 7 Frame used in butt welding 1.2-in. plates under restraint. Bierett and Albers[6]

Dimensions shown are in millimeters. Gage length l_1/l_2 used for shrinkage measurements. Gage lengths l_3/l_4 and l_5/l_6 used for shrinkage stress measurements

The authors conclude that, for a given weld cross section in plates of considerable thickness, the empirical formula holds for multiple-layer, K-shaped, horizontal welds, whereas shrinkages 20 to 25% greater than calculated must be expected in the case of double V, vertical welds. However, no account is taken of the effect produced by the relative positioning of the plates during welding. A more conclusive proof of the advantage of a K-shaped joint consisting of a great number of layers over a double V-joint welded with only a few layers would have been achieved, had both types been compared in both the horizontal and the vertical position.

Bierett and Albers[6] also investigated transverse shrinkage and the effect of interpass temperature in heavy steel plate welded by multipass sequence under heavy restraint. The plates, made of St 52 structural steel and measuring 1.2 x 20 x 12 in., were butt welded along the shorter edge while clamped between two heavy plates. The effective length of the restraint could be varied by means of a system of tapered pins and holes drilled in the heavy plates and specimens, Fig. 7. Coated Kjellberg St 52A electrodes of 0.16 in. and 0.20 in. diameter, and 170 and 220 amp. d. c., respectively, were used. A total of four tests was reported on: (1) Moderate restraint; 72° double V-weld in 16 passes, without interpass cooling. (2) Same as test 1, but under rigid restraint. (3) Rigid restraint; U-weld with 1.2 in. maximum joint width and 22 passes and 2 root passes; interpass cooling of about 20 min. each after layers 1, 2, 3, 4, 6, 9, 12, 14, 17, 19 and 22. (4) Same as test 3, but without interpass cooling. The root spacing was 1/8 in., and all passes were continuous across the entire

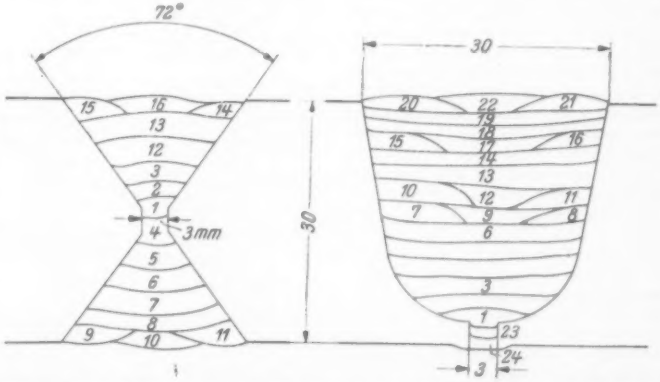


Fig. 8 Types of joints and welding sequence used in butt welding 1.2-in. plates under restraint. Bierett and Albers[6]

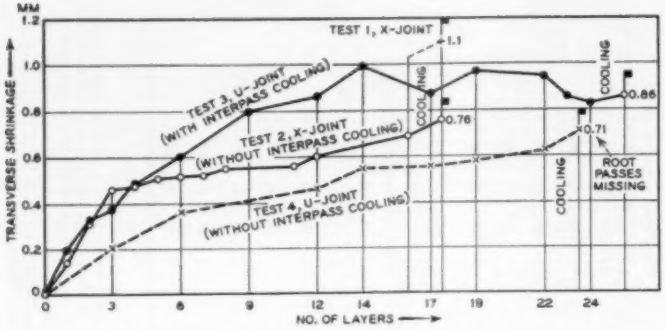


Fig. 9 Average transverse shrinkage for X- and U-welded 1.2-in. plate specimens under restraint. Bierett and Albers[6]

Measurements l_1/l_2 over 4-in. gage length

length of the joint, with the direction of welding reversed after each pass. The joint dimension and layer sequence are shown in Fig. 8.

Shrinkage measurements were made over a 4-in. gage length (l_1/l_2) across the joint by means of Siebel-Pfender type expansion gages, after releasing the specimens from their restraint. Figure 9 shows the transverse shrinkage as a function of the number of layers for tests 2, 3 and 4, and gives the total shrinkage for test 1. The shrinkage values for the finished and cooled welds were: for test 1, 0.043 in.; for test 2, 0.030 in.; for test 3, 0.034 in. and for test 4, 0.028 in. These results merely prove that the degree of restraint greatly affects the total shrinkage, but the curves in Fig. 10 clearly show the path taken by shrinkage during the welding. It is interesting to note that most of the transverse shrinkage in thick plates occurs long before the weld is finished, that is during the earlier passes.

While in Fig. 9 the mean values of the shrinkage are shown, Fig. 10 represents a plot of the individual shrinkages of the face (l_1) and root (l_2) sides of the specimens during tests 3 and 4 (U-welds). From this it can be seen that, while the continuously welded specimen shows uniform shrinkage on both sides throughout the major part of the welding, the specimen welded with cooling between passes not only undergoes a greater shrinkage, but also tends toward angular distortion starting with the first pass. On the basis of their experiments, the authors recommend continuous welding without interpass cooling, especially during the early passes in thick plates, in order to avoid unfavorable bending effects, such as angular distortion or bending stresses. They also emphasize the importance of limiting the weld cross section to a minimum. However, they do not see any particular advantage in using a double V in preference to a single U-weld.

A mathematical approach for determining transverse shrinkage in unrestrained butt-welded plates is also proposed by Oestreicher.[7] Thus $S_t = (t_m - t_f) \cdot \alpha \cdot W$, where S_t is the transverse shrinkage in in., t_m is the mean temperature between 1100° F. and room temperature arrived at by plotting the temperature distribution over the entire width of the plates when the weld has cooled to 1100° F., since shrinkage does not become effective until the specimen has cooled to about 1100° or below; t_f is the final temperature; α is the linear coefficient of expansion corrected for the mean temperature; W is the width of one plate.

The author claims that this formula has been proved by laboratory experiments, but he does not indicate what edge preparation, plate thicknesses and other welding conditions were employed. Neither does he discuss a case where the plates were of unequal widths.

Longitudinal Shrinkage

While longitudinal shrinkage, that is shrinkage parallel to the weld, plays an important role in the stress picture of a welded structure,[8] most of the information found on longitudinal shrinkage refers to fillet welds and weld beads. These are discussed in other sections of this Review.

An empirical formula for estimating the contraction along a welded butt joint is given by King:[9]

$$\frac{0.12 \times I \times L}{100{,}000 \times t},$$

where I is the current in amp., L is the length of the weld in in. and t is the plate thickness in inches. For multiple pass welds he recommends that $^1/_{10}$th of the contraction for the first pass be added for each successive pass. No details are given.

FILLET WELDS

Transverse Shrinkage

As for butt welds, the transverse shrinkage of fillet welds is composed of two parts: (*a*) the shrinkage of the deposited metal itself, and (*b*) the shrinkage of the parts subjected to unequal heating during welding. Because of the great number of different variables involved, however, no one formula or shrinkage chart could cover all situations.

Guyot[3] conducted an extensive investigation of the transverse shrinkage for which he used a cross-shaped assembly, Fig. 11 (*a*) containing stiffeners and end plates, welded with double fillet welds of four layers each, full weave, in the flat position. The three assemblies studied consisted of (1) plates 0.32 in. thick, (2) plates 0.63 in. thick and (3) plates 0.95 in. thick. Covered electrodes were used; for the first layer 0.128 in. diameter, 135 amp., for successive layers 0.16 in. diameter, 180 amp.

Shrinkage was measured between the end plates at five different points. A measurement was made after depositing each of the four layers comprising the fillet welds in welding the two end plates and three groups of four stiffener plates to the cross-shaped base. Thus, a total of 960 measurements was taken on the three assemblies and plotted against the cross sectional area of the weld, shown in Fig. 11

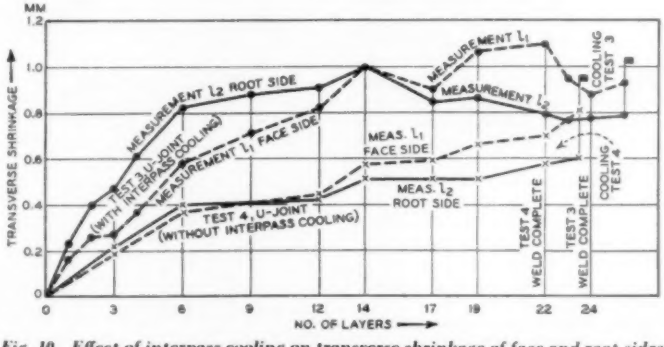


Fig. 10 Effect of interpass cooling on transverse shrinkage of face and root sides for U-welded 1.2-in. plate. Bierett and Albers[6]

Measurements l_1 and l_2 over 4-in. gage length

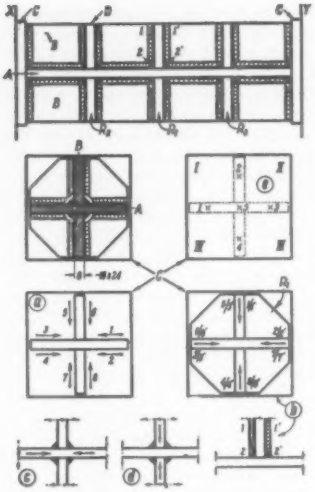


Fig. 11 (a) Cross-shaped assembly with stiffeners and abutment plates to measure transverse shrinkage due to welding the stiffeners and abutment plates

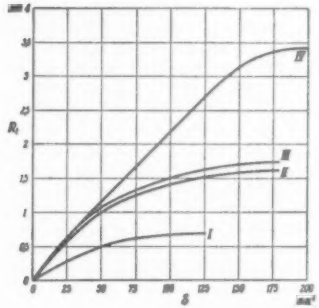


Fig. 11 (b) Results of transverse shrinkage measurements on welded stiffeners and abutment plates

R_t = transverse shrinkage, mm.; S = cross sectional area of weld, sq. mm.

(*b*). Curve I represents the transverse shrinkage in the through plate due to the welding of one group of stiffeners. It reaches a maximum value of 0.65 mm. (0.026 in.) for a cross sectional area of the fillet equal to 75 mm.^{2} (0.116 sq. in.), corresponding to an 8.5 mm. ($^3/_8$ in.) fillet. Curves II and III represent the shrinkages after welding the second and third group of stiffeners to the same through plate, and Curve IV shows the change in transverse shrinkage as a function of the increasing cross section of the fillet welds of the two end plates.

In his conclusions, Guyot finds no systematic difference in shrinkage with the three plate thicknesses. He explains that the marked increase in shrinkage after welding the second group of stiffener plates is due to the partial release of internal

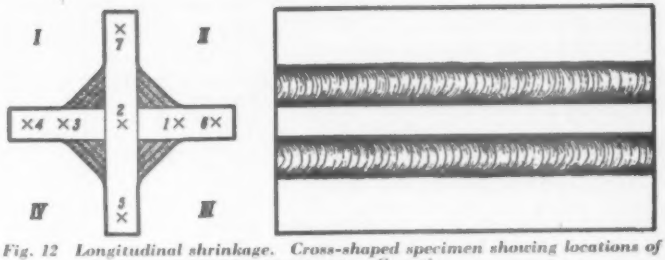


Fig. 12 Longitudinal shrinkage. Cross-shaped specimen showing locations of measurements. Guyot[3]

stresses caused by the first pair of plates, whereas the small increase after the third group is due to the fact that the internal stresses are so distributed in the through plate as to oppose new shrinkage forces. Similarly, a fourth group of stiffeners welded to the same through plate would cause no further increase. Welding a perpendicular plate to a through plate by means of fillet welds causes a transverse shrinkage in the plane of the perpendicular plate which is almost proportional to the area of the fillets. Maximum shrinkage of 3.5 mm. (0.138 in.) is attained for a weld cross section of about 200 mm.^{2} (0.31 sq. in.), corresponding to a 14 mm. (0.55 in.) fillet. A comparison of Curve IV with curves I to III shows that for a given fillet size the perpendicular plate suffers more shrinkage than the through plate.

Longitudinal Shrinkage

Guyot[3] also made an extensive study of longitudinal shrinkage in fillet welds. As in the case of transverse welds, he found that longitudinal shrinkage is primarily a function of the total cross section of the joints involved. However, in contrast to transverse shrinkage, where a portion of the shrinkage is composed of the contraction of the deposited metal as well as the permanent contraction of the base metal heated nonuniformly during the welding, here these two effects are not additive, but act in opposite directions and thus restrain each other. This restraint will be the more effective, the thicker and wider are the plates. He therefore calls "resisting cross section" the total cross section of the welded plates in the transverse direction.

In order to maintain symmetry in the transverse section and thereby keep longitudinal deflection to a minimum, Guyot used for his experiments eight cross-shaped assemblies consisting of plates of increasing widths and thicknesses, thus obtaining resisting cross sections varying from 0.4 to 18.2 sq. in., Fig. 12 and Table 2. Each fillet was made in four layers, and longitudinal shrinkage was measured after the deposition of every corresponding pair of layers in the opposite angles I and III, and II and IV. In this manner seven measurements, indicated in Fig. 12, were taken after every pair of layers, amounting to a total of 448 shrinkage measurements which are shown plotted as a series of curves in Fig. 13. Each curve represents the variation of longitudinal shrinkage per inch of length of weld as a function of the resisting cross section of the assembly and for a constant weld cross section. The average cross section of the weld metal deposited was computed from the difference in weight of the bare core wire of the electrodes before and after welding. No welding details are given other than that 0.16 in. diameter covered electrodes were used, but it may be assumed that the same procedure was used as in the tests for transverse shrinkage.

The exponential appearance of the curves in Fig. 13 shows that the shrinkage, which attains high values for small resisting cross sections, falls rapidly as the section increases. The shrinkage tends to become stabilized when the resisting cross section of an assembly exceeds a certain value, as indicated by the dotted curve in Fig. 13. Because of the wide variation in longitudinal shrinkage, the author cautions that it is very essential to evaluate the resisting cross section precisely, when computing the shrinkage a priori.

The cross-shaped assemblies used to determine the curves in Fig. 13 were all 500

Table 2—Transverse Dimensions of the Eight Assemblies and Their Resisting Cross Section. Guyot[3]

Thickness		*Plate width*		*Resisting cross section*	
Mm.	*In.*	*Mm.*	*In.*	*Sq. mm.*	*Sq. in.*
4	0.16	34	1.34	256	0.40
10	0.39	40	1.57	700	1.08
16	0.63	55	2.16	1,560	2.4
16	0.63	71.5	2.81	2,030	3.2
16	0.63	105	4.1	3,104	4.8
16	0.63	195	7.7	5,984	9.3
16	0.63	275	10.8	8,550	13.3
16	0.63	375	14.8	11,774	18.2

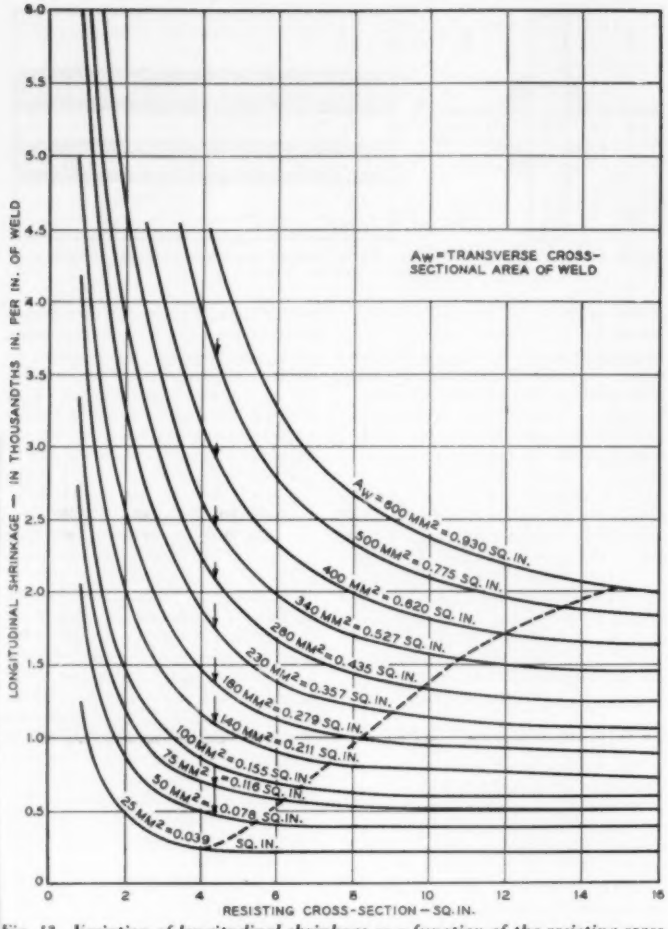
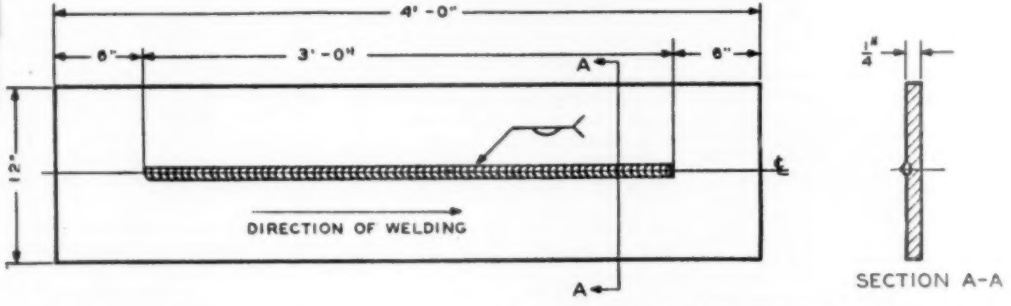

LONGITUDINAL SHRINKAGE — IN THOUSANDTHS IN. PER IN. OF WELD
RESISTING CROSS-SECTION — SQ. IN.
A_W = TRANSVERSE CROSS-SECTIONAL AREA OF WELD
A_W = 800 MM² = 0.930 SQ. IN.
500 MM² = 0.775 SQ. IN.
400 MM² = 0.620 SQ. IN.
340 MM² = 0.527 SQ. IN.
280 MM² = 0.435 SQ. IN.
230 MM² = 0.357 SQ. IN.
180 MM² = 0.279 SQ. IN.
140 MM² = 0.211 SQ. IN.
100 MM² = 0.155 SQ. IN.
75 MM² = 0.116 SQ. IN.
50 MM² = 0.078 SQ. IN.
25 MM² = 0.039 SQ. IN.

Fig. 13 Variation of longitudinal shrinkage as a function of the resisting cross section for 12 different values of weld cross section. Guyot[3]

mm. (19.7 in.) long. In order to study the effect of weld length on longitudinal shrinkage per unit length, Guyot repeated his experiments on a cross-shaped assembly four times as long, with a resisting cross section of 4.3 sq. in. (2800 mm.²). He also varied the sequence of depositing layers in the two pairs of angles slightly to keep the longitudinal deformation of the assembly to a minimum. In addition to measuring the total shrinkage at five different points on the ends of the cross, shrinkage was also measured between eight points along the four longitudinal edges of the plates. The averages of these measurements, carried over to Fig. 13, are represented by the vertical passing through the resisting cross section at a value of 4.3 sq. ins., and the differences between the measurements and the curves are shown by arrows. These differences vary from 0 to 0.008 in. as the cross-sectional area of the weld decreases from large to small values, being of the same order of magnitude as the dispersion observed in separate tests. Consequently, Guyot concludes, the curves of Fig. 13 apply to long as well as short welds.

Moon[10] reports on longitudinal shrinkage in fillet welds of very great lengths. In a fillet weld 180 ft. long in $^3/_{16}$ in. thick plate, welded in a single pass, a shrinkage of 2 in. was observed in the length of the seam. On the other hand, in a fillet weld 120 ft. long in $^3/_8$ in. plate using three passes, no shrinkage occurred. Each run of weld was of the same size as in the previous case. From that he concludes that, if the size of the run of weld is small in comparison with the thickness of the work, little or no distortion takes place along the weld and the shrinkage is balanced by stretching in the weld metal. Unfortunately, the width of the plates and the size of the fillet are not given, nor is it stated whether the plates were restrained against warping. However, considering the length



Welding conditions				Plate movement			
Electrode size, in.	*Current amp.*	*Voltage*	*Speed of welding, in./min.*	*Total expansion, in.*	*Total contraction, in.*	*Net total contraction from original length, in.*	*Unit contraction from original length, in./ft.*
$^1/_8$	100	25	6	0.008	0.0215	0.0135	0.0045
$^3/_{16}$	170	27	$5^3/_4$	0.0217	0.0430	0.0213	0.0071
$^1/_4$	200	30	5	0.024	0.0518	0.0278	0.0093

Fig. 14 Details of longitudinal weld-bead shrinkage test specimen—results of tests. Bibber[11]

of the weld in the first example, the total shrinkage of 2 in. converted to unit shrinkage amounts to but $^1/_{1000}$ in. per in. of weld, which is well within the lower range of shrinkage values of Guyot's chart, Fig. 13.

On the subject of size of deposit, Moon modifies his previous conclusion somewhat by stating that a comparatively large run of metal may be deposited from a large gage electrode without causing distortion if the speed of travel in making the weld is sufficiently fast.

Weld Beads

Bibber[11] demonstrates the shrinkage effect of longitudinal weld beads of various cross section deposited on the flat side of a steel plate. A plate, $^1/_4$ in. thick, 1 ft. wide and 4 ft. long was welded at one end to a heavy slab. The free end was held down by a strap in such a way that the plate could elongate or contract lengthwise, but could not curl or lift from the base. A dial gage was placed in contact with the free end of the plate. Beads of welding 3 ft. long were deposited on top of the plate, using various electrode sizes and heat inputs. The physical setup, the welding conditions and the results of the tests are shown in Fig. 14. As the heat input increased, the longitudinal contraction increased also. In fact, the data when plotted show that the unit net shrinkage increased in direct proportion to the unit heat input.

The deformation due to shrinkage in unsymmetrically welded steel plates was studied by Malisius.[12] The effects of gas and electric arc welding were also compared. The test pieces consisted of mild carbon-steel plates, 0.59 in. thick, 27.6 in. long and of various widths. The unsymmetrical weld effect was produced by depositing weld beads of 0.03 sq. in. cross section along one longitudinal edge. The cross section of the weld deposit was carefully maintained constant for all specimens tested. The pieces were stress relieved before the welding by heating to 1200° F. and cooling in the furnace. Five plates, varying in width from 1 to 4 in., were welded by gas, using a welding speed of $2^3/_4$ in. per minute. For the electric arc welding, three plates, 1.4, 2.6 and 4 in. wide, were tested. The welding data are: coated electrodes, 0.160 in. diameter; 180 amp., 32 v., d. c.; welding speed $7^1/_2$ in. per minute.

To measure the deformation caused by the welding, the specimens were placed edgewise on two fixed posts, $23^1/_2$ in. apart, before and after the welding, and the maximum permanent deflection of the unwelded edge measured from the horizontal. These deflections are shown in Table 3.

Arc welding caused the specimens to deform more than gas welding, although the welding speed was higher and the heat-affected zone smaller, which is contrary to normal expectations. As the author explains, for certain cross sections the upset area may be larger though the heat-affected zone be smaller, since the remaining cross section, being larger, is more effective in preventing the heat from spreading out. He cautions, however, that this condition does not apply in the case of larger plate cross sections. The deformation tends to become stabilized for plates over $2^1/_4$ in. wide which he ascribes to the internal stress conditions in the plates, that is, an equilibrium exists between the tensile stresses in the upset portion and the bending stresses in the remaining portion. In the smaller plate widths the influence of the upset and shrinkage apparently exceeds that of the bending stresses. Malisius concludes that no simple relationship exists between the deformation and the cross section or the moment of inertia of the plates.

Table 3—Deformation of Unsymmetrically Welded Steel Plates of Various Widths. Malisius[12]

Width of plate, in.	*Maximum longitudinal deflection, in.*
0.95	0.190
1.4*	0.160
1.5	0.145
2.3	0.040
2.4*	0.090
2.8	0.040
4.0	0.040
4.0*	0.055

* Welded with the electric arc. All others welded with gas.

ANGULAR DISTORTION

Angular distortion, that is the lifting of the free ends of the two members being joined out of their initial plane, with the weld as the axis of rotation, is a result of unsymmetrical shrinkage forces acting across the width and/or through the thickness of the members at the joint. The advantage of using larger electrodes and fewer runs in preventing angular distortion has been fairly definitely established.[1, 2, 13] Johnson[14] and Heigh[15] illustrate this further for butt welds, Fig. 15, and fillet welds, Fig. 16, by comparing the angular distortion produced in the same size weld for various electrode sizes and numbers of passes required to complete the weld. The experiments were carried out in such a manner that the plates were free to tilt, and the measurements were taken at an unspecified radius. The economical advantage in employing larger electrodes, when angular distortion is the main concern, is therefore also proved. To reduce angular distortion even further, Johnson recommends presetting of the plates against the anticipated distortion or a preventive support, the latter method being the more economical and practical.

Campus[4] also investigated the effects of the number of passes, the interpass temperature and the length of the weld, on the magnitude of the angular distortion in butt-welded 0.55 in. plate. The results are shown in Tables 4, 5 and 6, respectively. A $^1/_{16}$-in. root gap was used in each case.

Table 4—Effect of Number of Passes on Angular Distortion in 90° V Butt Joints. Campus[4]

Number of passes	*Angular distortion, degree*
7	2° 50′
9	6° 15′
10	7° 30′
12	8° 30′
15	11°

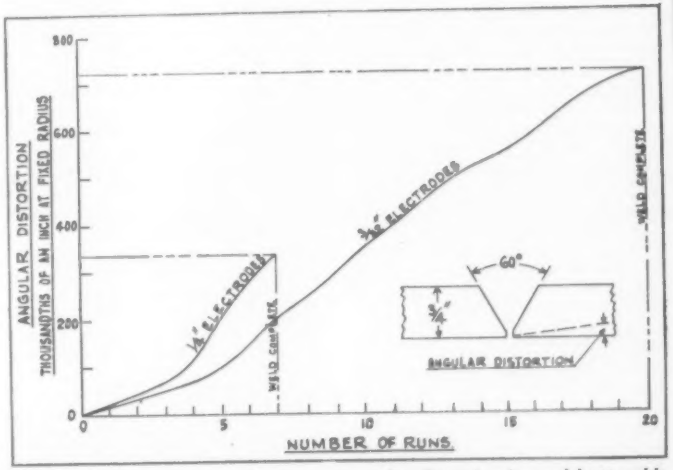


Fig. 15 Effect of number of passes on angular distortion in veed butt welds. Johnson[14] and Heigh[15]

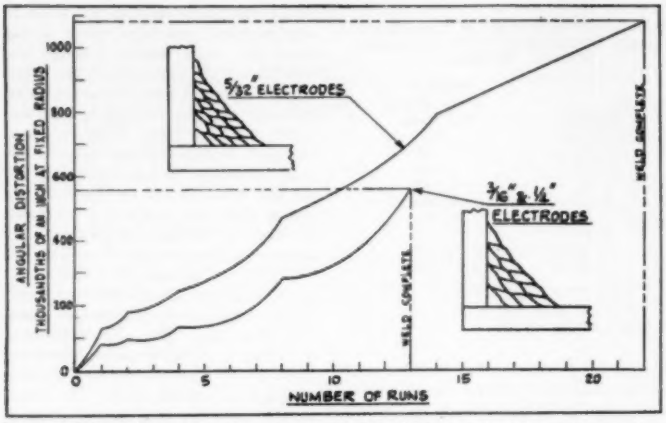


Fig. 16 Effect of number of passes on angular distortion in fillet welds. Johnson[14] and Heigh[15]

Table 5—Effect of Interpass Temperature on Angular Distortion in 80° V Butt Joints. Campus[4]

Procedure	*Angular distortion degree*
Passes in continuous succession	4° 55′
Interpass temperature = 300° F.	4° 35′
Interpass temperature = 70° F.	8° 20′

Table 6—Effect of Length of Joint on Angular Distortion in 80° V Butt Joints. Campus[4]

Length of joint, in.	*Angular distortion,* °
4	2° 40′
6	10° 49′

Campus' results were obtained from unrestrained specimens. No welding details are given other than that 0.160 coated electrodes were used. However, the results serve to bring out the relative importance of some of the influencing factors involved in producing a weld with a minimum of distortion. His findings regarding the increasing angular distortion with lower interpass temperatures tend to confirm those of Bierett and Albers.[6]

In regard to angular distortion in butt welds and its control by suitable joint design, the British FE. 7 Committee on Design of Welded Machinery Construction and Parts[16] makes the following recommendations: For plates 5/16 to 5/8 in. thick, single V-joints may be used, provided the joint is not very long. For plates over 5/8 in. thick, the single V is not advisable other than for joining narrow bars or when turning over is inconvenient. For plates over 3/4 in. thick, a double V, or a single or double U-joint may be used, both U-types of joint becoming more economical as compared to double V-joints with increasing plate thickness, since the savings in welding costs will more than offset the increased preparation costs.

COMPARISON BETWEEN MANUAL AND AUTOMATIC WELDING

A comparison of the shrinkage effects of manual welding with automatic submerged arc welding is offered by Rybakov.[17] Transverse and longitudinal shrinkage measurements in butt-welded plates, measuring 0.4 x 19 1/2 x 4 1/2 in., are shown. The automatic welding was done in the form of a single pass, after a manual root weld was laid in a 60° groove to a depth of about 0.16 in. from the apex side. The manual welding was effected in a single pass, in a 60° V-groove, with a root face of about 0.06 in. and an 0.08-in. root clearance. Prior to the welding of the butt joint, the plates were tacked together at three points along the longer edge, i.e., at a distance of 2 in. from each end and in the middle. Longitudinal as well as transverse shrinkage was measured with a portable strain gage equipped with a dial indicator between points marked off on the transverse and the longitudinal edges of the specimens, before and after the welding.

Further welding details and shrinkage curves are shown in Figs. 17 and 18. The results indicate that single pass, automatic, submerged arc welding of butt joints, preceded by manual root welding, gives an average distortion that is 30 to 40% lower than that in single pass manual welding over the entire thickness of the plates. The author attributes the difference between manual and automatic welding with respect to the distortion produced to the difference in the fusing power of these processes and to the different magnitudes of thermal energy transferred during the welding. The effective thermal energy was 4750 cal. per cm.^{3} in the case of automatic, and 5825 cal. per cm.^{3} in the case of manual welding, which represents a difference of about 20%.

Rybakov also measured the permanent deflection in a 20 in. long T-section, formed by two 0.47 x 4.3-in. steel plates fillet welded by both automatic and manual welds in a single pass on both sides of the web without cooling. In contrast to the results obtained with butt welds, the maximum deflection of the flange was nearly twice as great for the submerged-arc as for the manual method, 0.043 and 0.025 in., respectively. This is in spite of the fact that in the automatic welding less filler metal (280 vs. 430 gm.) was deposited and less thermal energy (4240 vs. 5500 cal. per cm.^{3}) was used.

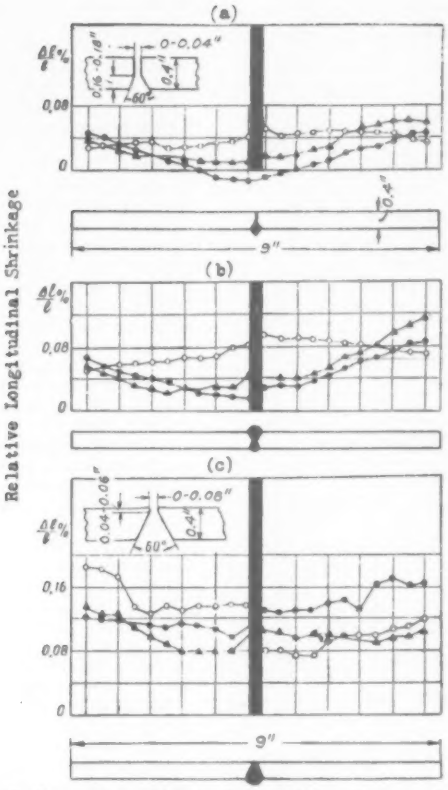
According to the author, this inconsistency is due to the difference in the depth to which the flange and web are fused under the two methods. In automatic welding, the center of gravity of the shrinkage zone is shifted from the center of gravity of the composite section in the direction of the flange for a greater distance than in manual welding.

METHODS OF CONTROLLING DISTORTION DUE TO WELDING

Numerous papers have been written containing general rules and recommendations of how contraction, distortion and warping due to shrinkage in welding can effectively be prevented, or at least reduced.

Typical of those is a set of recommendations by Moon.[10]

1. Adjust design so that the pull of one weld may be counteracted by the pull of another weld.
2. Employ rigid tacking at intervals depending on the thickness of the plates.
3. The individual runs should be made small in comparison with the thickness of the plate.
4. The speed of travel during welding should be as rapid as possible to prevent the heat from running ahead of the arc.
5. The welding may be done in small sections to spread the heat evenly about the work, if necessary using the "back stepping method" or some variation of it.
6. The shape of the piece should be checked from time to time during welding, and, if distortion commences, work should be stopped on the weld causing deformation and welding started on the opposing weld to bring the piece back straight.
7. If the structure does not permit balancing the welding, the welds may be peened after each run to stretch the weld metal and thus to counteract the cooling shrinkage.



To Figs. 17 (*a*) and 18 (*a*)

Symbols	Welding variables: Welding speed M./hr.	Welding variables: Welding speed Ft./hr.	Welding variables: Current, amp.	Welding variables: Voltage, v.	Welding variables: Electrodes	Metal deposited, gm.
●	6	20	200	26	OMM-5, ø 4 mm., 0.16 in.	110
○	6	20	200	26	OMM-5, ø 4 mm., 0.16 in.	130
▲	6	20	200	26	OMM-5, ø 4 mm., 0.16 in.	100

To Figs. 17 (*b*) and 18 (*b*)

Symbols	Welding variables: Speed M./hr.	Welding variables: Speed Ft./hr.	Welding variables: Current, amp.	Welding variables: Voltage, v.	Welding variables: Type of flux	Metal deposited, gm.
●	36	118	800–850	37–40	OSTs –45	263
○	36	118	800–850	37–40	OSTs –45	270
▲	42	138	800–850	35–37	OSTs –45	290

To Figs. 17 (*c*) and 18 (*c*)

Symbols	Welding variables: Speed M./hr.	Welding variables: Speed Ft./hr.	Welding variables: Current, amp.	Welding variables: Voltage, v.	Welding variables: Electrodes	Metal deposited, gm.
●	7½	25	260	25	OMM-5, ø 5 mm., 0.2 in.	280
○	7½	25	260	25	OMM-5, ø 5 mm., 0.2 in.	160
▲	7½	25	260	25	OMM-5, ø 5 mm., 0.2 in.	205

Fig 17 Comparison of longitudinal shrinkage in manual and submerged arc-welded plate. Rybakov[17]

(*a*) Manual backing weld prior to automatic weld; (*b*) after backing and automatic welding; (*c*) manual weld.

Among the most frequently listed general recommendations in the literature is one concerning the number and size of passes. Thus, Cellers[18] writes with reference to transverse shrinkage "Single pass welds should be used whenever possible, as they result in only one shrinkage."

Similarly, Gill[19] writes: "It is not good practice to use two passes of welding when one pass will produce a satisfactory joint, since the reheating of the parent metal during multipass welding will produce additional shrinkage and cause increased distortion." Presumably, although not stated, Gill compares one large pass with two or more smaller ones of an equivalent total weld cross section. Also not stated in many of these general statements is the thickness of material and the type of shrinkage referred to by the author—transverse or longitudinal.

Molloy[20] and Morrill[21] (longitudinal shrinkage) also advocate the use of fewer passes to reduce the amount of shrinkage and warping. Hadjispyrou[22] expresses the same opinion by recommending a high rate of heat input and large electrodes and welding currents, when transverse shrinkage or angular distortion are objectionable. On the same subject, Hill[23] recommends greater welding speed. For the reduction of distortion in large welds, Brooking[24] advises the use of small passes, or moderately large passes with peening between passes.

A solution to the problem of the optimum of layers for controlling transverse shrinkage for a given size of weld is given by Guyot,[3] who writes: "The total shrinkage obtained with several layers is always less than that for the same joint made in a single layer. To reduce shrinkage to a minimum we should endeavor to weld the joint with a maximum number of layers. That is, we should use small electrodes.

"It seems that speed of welding has a great effect on shrinkage, particularly indirect shrinkage (the contraction of the joint at a point which is caused by the shrinkage of the preceding parts of the joint). Since indirect shrinkage arises from the fact that the entire joint cannot be made simultaneously, it appears that the joint should be made as fast as possible. This implies high currents with large electrodes. Our preceding paragraph, however, called for the use of small electrodes. A compromise therefore is in order: we should use electrodes of medium diameter."

As for controlling shrinkage and shrinkage effects by directional welding, Molloy[20] and Gill[19] both recommend welding from the point of higher restraint to the point of lower restraint. Similarly, when welding a rigid unit, Cellers[18] advocates the welding to be carried out from the center toward the free edges.

The advantages of directional welding in preventing or controlling distortion and residual stresses are demonstrated by Holt[25] in reporting on a series of tests conducted at the University of Washington. The results showed that, in butt welding tacked steel plates, a considerable reduction in transverse shrinkage as well as angular rotation in the plane of the plates was brought about by welding both from the center toward the edges and from the edges toward the center, as compared with continuous unidirectional welding.

The specimens used consisted of two mild-steel plates, 3 in. wide by $^3/_{16}$ in. thick and over 5 in. long. The edges were beveled 30° each and the plates were butt welded from one side, without root open-

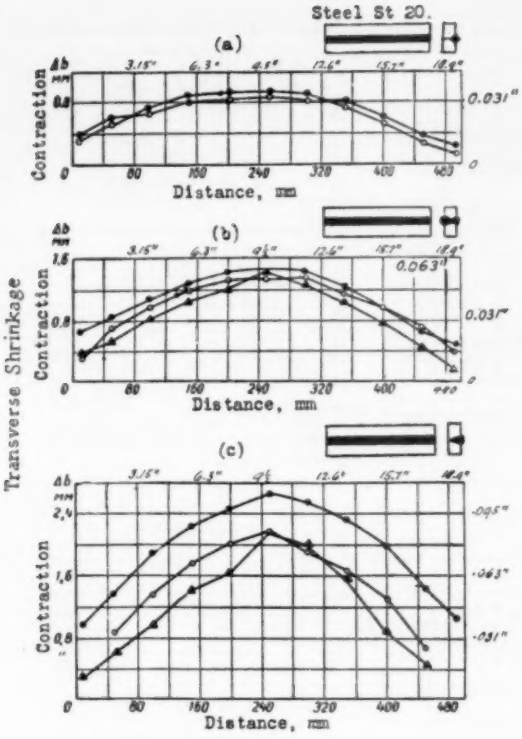


Fig. 18 Comparison of transverse shrinkage in manual and submerged arc-welded plate. Rybakov[17]

(a) Manual backing weld prior to automatic weld; (b) after backing and automatic welding; (c) manual weld

ing. Speed of arc travel was 5 in. per minute, the electrodes were $^{5}/_{32}$ in. with 160 amp. d. c.

When the specimen was welded continuously all the way across, there was a transverse shrinkage of approximately 0.02 in. and 0.011 in. movement toward the weld one-half in. from the starting edge. One-half in. in from the finishing edge there was a movement toward the weld of 0.021 in. When the welds were made from the center of the specimen to the edges, there was a movement edgewise (transverse shrinkage) of only 0.004 in. toward the last weld made, a movement transverse to the weld of 0.0146 in. one-half in. in from the starting edge and 0.0198 in. in from the finishing edge. When the sequence was changed and the direction of weld travel was from the edges toward the center of the specimen, the corresponding measurements were 0.01 in., 0.0129 in. and 0.0185 in.

A general treatment of the basic principles and control procedures of shrinkage in welding is presented by Aitchison.[26] Practical cases are discussed in the light of theory and simple examples, but no quantitative data are given. Similar information can be found in a paper by Jefferson.[27]

Bainbridge[28] treats the shrinkage problem in the gas welding of sheet metals. Since the volume of a sheet metal weld is small in comparison to that of heated base metal, transverse shrinkage can usually be ignored, but on long welds longitudinal shrinkage may be a considerable contribution to buckling along the joint. The chief cause of buckling can be minimized by confining the welding heat to a narrow zone. Here a properly designed jig can serve in dual capacity, (1) to maintain alignment and restrain movement of the metal, and (2) to act as a chill or by-pass for the excess heat. Joints should be so designed as to require the least amount of heat, the square edge butt and the square flange butt being the most preferred. Pauses during welding tend to increase distortion, therefore each weld should be carried through without interruptions. Tacks to hold parallel joints in alignment should be spaced more closely for thin than for thick metals, the recommended average for steel being 4 to 6 in. apart.

Bibber[11] shows how shrinkage in the welded erection of large plate structures can be constantly corrected and allowed for as the work proceeds. For instance, in the case of ships, a corrective welding sequence, starting at the mid-section and progressing forward and aft with upper welding preceding the lower, succeeded in preventing the hull from lifting from its keel blocks. With the previous method, wherein the welding of the bottom areas had been largely completed before the welding of the upper section, a lift of the bow and stern from its blocks of as much as 8 in. had been experienced.

A method of controlling distortion in joining sections of huge steel floating dry docks, used by the Bureau of Yards and Docks on 77 floating dry docks, is described by Dawson.[29] The method consisted of taking daily elevations on the pontoon deck, at the same time of the day in order to minimize atmospheric influences. The deck was divided into 18 spaces longitudinally, and plus and minus deviations from the transverse dock centerline elevation measured. The following day's welding operations were then adjusted so as to counteract any distortions. This could usually be accomplished by judicious positioning of the welders, but also by varying the welding speed and sequence. At one location a three-sectional 528 ft. long dock was constructed in this fashion with only a $^{3}/_{8}$-in. deviation.

A skip procedure and multidirectional welding sequence, which were used in welding a cartridge case ejector chute to keep shrinkage distortion to a minimum, are described by Henderson.[30] The chutes were made of $^{1}/_{8}$-in. high tensile steel plate, to which stainless steel rails, $^{3}/_{4}$ by 1 in. were welded with stainless and carbon steel rods.

Farmer[31] tells of a method of keeping distortion down to very small tolerances in single pass untacked welds on aircraft engine mounts and support bays. This was done by means of welding fixtures mounted on trunnion stands for positioning so that all welds could be made down hand. The procedure used in welding a large (46 x 20 ft.), flat brewery tank is reported.[32] The material consisted of 10 gage, type 302 stainless steel, the process used was the submerged automatic arc with a $^{5}/_{32}$-in. electrode, and the maximum permissible distortion was $^{3}/_{16}$ in. from flat for the bottom of the tank. A jack clamp used for prebending the edges of $1^{1}/_{2}$-in. armor plate to compensate for shrinkage distortion in preparation for butt welding by a multipass process is described by Wright.[33]

SUGGESTED RESEARCH TOPICS

This Review of the Literature would not be complete, did it not include suggestions for additional research. The topics below

have been selected, because the authors, in reviewing reports of past investigations, found that they have one or more of these characteristics: (1) lack of information, (2) contradictory information, (3) information qualitative, but not quantitative and (4) importance with respect to a better understanding.

The following list of topics for research on Shrinkage Distortion is therefore suggested:

1. The effect of the number of passes on total transverse shrinkage in butt welds of a given size and type of joint. This should include a study of the effect of the type of deposit and of the interpass temperature.
2. An investigation of all aspects of shrinkage distortion in the welding of nonferrous metals, such as copper, aluminum, nickel, *et al.*
3. A comparison of the shrinkage effects in automatic as against manual welding.
4. Studies of shrinkage phenomena in other types of welding, such as the inert-gas-shielded electric arc and seam welding.

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Comment by H. Weisberg

We have recently collected some data on this subject in connection with the field rewelding of Schedule 160, A.S.T.M. Spec. A158-36, Grade P-1, carbon-moly steel piping following our investigation of graphitized welded joints.

The welding grooves were prepared to the A.S.A. 30° standard bevel using a $^1/_8$-in. thick backing ring and allowing a $^3/_{16}$-in. gap before welding. The joints were preheated at 600° F. before welding and given a 1300° F., 4 hr., postheat treatment immediately following welding. The shrinkages following the completion of the weld, taken by an 8-in. tram measurement across the weld are as follows:

8-in. pipe—of 14 joints checked, 11 showed $^1/_8$-in. shrinkage, while one showed $^1/_{16}$ in. and 2 showed $^3/_{16}$-in. shrinkage.

12-in. pipe—of 21 joints checked, 14 showed $^1/_8$-in. shrinkage, while one showed $^1/_{16}$-in. shrinkage; 4 showed $^3/_{16}$; and 2 showed $^1/_4$-in. shrinkage.

A check of our welding data indicates that the largest shrinkage values were obtained in every case where a connection was made between the schedule 160 pipe and a heavy valve body. The results, however, are inconsistent because in several other cases the average shrinkage values also were obtained for similar conditions. The welding was all performed under the same procedure specifications and the same sizes of rods were used. Current input and preheat and postheat temperatures were controlled as close as is practicable. We are unable to offer any further explanation for the variation in shrinkage.

H. Weisberg is a Mechanical Engineer connected with Public Service Electric and Gas Company of New Jersey.

Welding Research Council Bulletins for Sale

E11.1 Steel Compositions and Specifications by Charles M. Parker. 1949. 11 pages 50c ☐

E11.2 The Nature of the Arc by J. D. Cobine. 1949. 6 pages 50c ☐

E11.3 Burn-off Characteristics of Steel Welding Electrodes by D. C. Martin, P. J. Rieppel and C. B. Voldrich. 1949. 8 pages 50c ☐

E11.4 Testing Pressure Vessels, by F. G. Tatnall, and Effect of Welding on Pressure Vessel Steels, by A. F. Scotchbrook, L. Eriv, R. D. Stout and B. G. Johnston. 1950. 12 pages 50c ☐

E11.5 How Plastic Deformation Influences Design and Forming of Metal Parts by John R. Low, Jr. Brittle Fracture in Mild Steel by J. S. Hoggart. 1950. 20 pages $1.00 ☐

AMERICAN WELDING SOCIETY

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Design Correlations for Cylindrical Pressure Vessels with Conical or Toriconical Heads

An investigation of head-to-shell junctures of pressure vessels. Comparison with measured and computed stresses

by H. L. O'Brien, E. Wetterstrom, M. G. Dykhuizen and R. G. Sturm

Abstract

This paper reports the second part of the results of an investigation of head-to-shell junctures of pressure vessels conducted at Purdue University. This was done under contract with the Design Division of the Pressure Vessel Research Committee of the Welding Research Council.

In particular, the secondary phase of the Purdue Project embodied tests on commercially fabricated vessels so designed as to correspond in a 6:1 ratio with two of the 8-in. diameter models previously tested and reported on. Curves are presented herein illustrating the comparison between measured and computed stresses for the 48-in. diameter vessels, along with conclusions.

Conclusions and an evaluation of the work to date are presented based on the findings from both model and full-scale tests.

IN ORDER to avoid repetition of the detail presented in a previous paper,* let it be said that, in general, the testing program for the large vessels was the same as that for the models. Although it was originally intended that the two 48-in. diameter vessels would be the full-size counterparts of Models 3 and 5, difficulties in fabrication unavoidably resulted in characteristics something less than ideal. The actual measurements of most critical interest are shown in Fig. 1. Strain gage locations were selected and the vessel wired for testing in essentially the same manner as previously outlined. Figure 2 shows the wiring for one of the large vessels and Fig. 3 is an over-all view of the same vessel as it was received from the fabricator.

ANALYSIS OF TEST DATA

Since the measured stresses were to be correlated with the theoretical stresses calculated by the approximate theory, the strain readings and corresponding pressures were plotted and the slopes of these pressure strain curves were determined graphically. The slopes obtained from these graphs are the stresses for 1 psi. of internal pressure. The graphical method for handling the measured strains was essentially the same for both the model tests and the tests of the 48-in. diameter vessels. This graphical method of handling results may be described in three steps: (1) The first and most important step was the plotting of measured strains as a function of the pressures at which they were obtained. These curves were all essentially straight lines for low pressures. (2) The second step was to determine the slope of the straight line portion of the pressure strain curves graphically and to evaluate the stress per pound of pressure by the application of the basic equations of the theory of elasticity to the measured values. (3) The third step was to correlate the measured strains for higher pressures with stress-strain curves modified for the biaxial state of stress existing at the gage. The correlation derived in step three proved very valuable in the analysis of residual stresses in the cone to shell juncture. The pressure strain curves were most valuable, however, in obtaining the measured values for use in comparison with the computed values.

RESULTS

48-In. Diameter Vessels—Materials

The detailed results of the supplementary program of tests on the materials used in the 48-in. vessels indicated that the materials were very uniform and isotropic and met the A.S.M.E. code specifications for A.S.T.M. A-201 firebox quality steel.

Theory vs. Test Results in the Elastic Range

Figure 1 represents the findings obtained from tests on the two 48-in. diameter vessels and, for convenience, also illustrates vessel A-1 designed with a sharp cone

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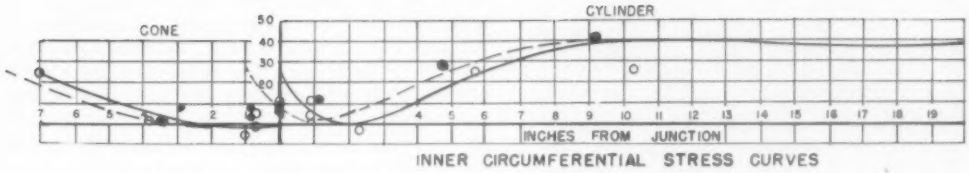
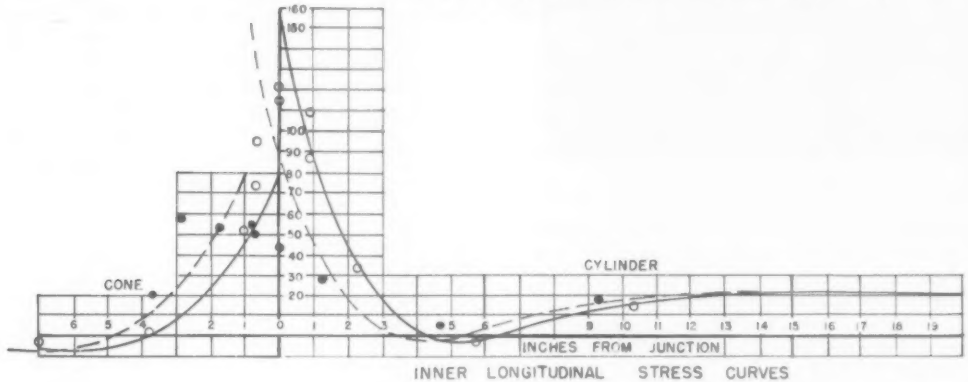
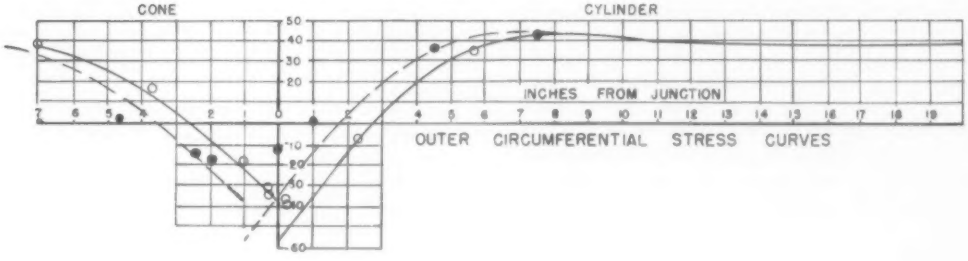
* R. G. Sturm, H. L. O'Brien, E. Wetterstrom, and J. Evans, "Stresses in Head-to-Shell Juncture of Pressure Vessels," THE WELDING JOURNAL, 29 6), Research Suppl., 285-s to 292-s (1950).



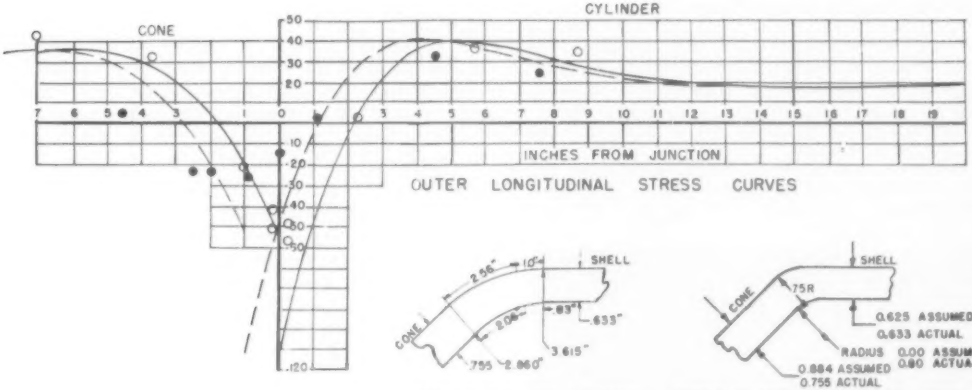
INNER CIRCUMFERENTIAL STRESS CURVES



INNER LONGITUDINAL STRESS CURVES

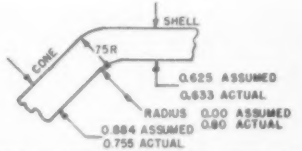


OUTER CIRCUMFERENTIAL STRESS CURVES



OUTER LONGITUDINAL STRESS CURVES

PROFILE AT CONE-SHELL JUNCTION
VESSEL A-2



PROFILE AT CONE-SHELL JUNCTION - VESSEL A-1

Fig. 1 Measured and computed stresses—48-in. vessel

Stress based on an internal pressure, P, of 1 psi. Theoretical Curves: Vessel A-1 solid curve; Vessel A-2 dashed curve. Experimental stress measurement: Vessel A-1, circles; Vessel A-2, solid circles

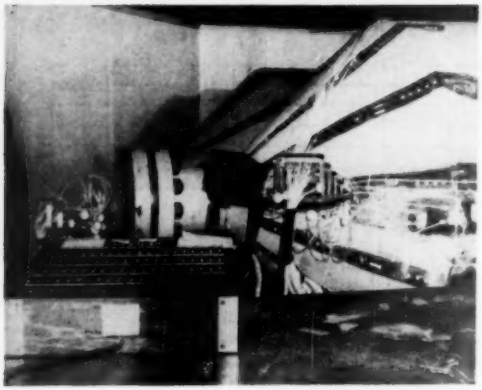


Fig. 2 Forty-eight inch diameter vessel wired for test

shell juncture and vessel A-2 embodying a mean knuckle radius of about 3 in. The resulting stresses have been plotted upon a set of theoretical curves which were based upon the Love-Meissner approximation for a sharp juncture. The origin for the calculation of the theoretical curves was arbitrarily defined as the junction of the center line of the head to the center line of the shell. In the case of the torus section of vessel A-2, the plane of intersection was chosen as the point of tangency between the torus and the shell. From a further theoretical consideration it appears that the maximum stress should occur near the center of the torus section. Recognizing this, the approximate theoretical curve was shifted toward the center of the torus, giving better agreement with the experimental stress distribution for vessel A-2. This shifting of the theoretical curve was not necessary to show agreement for vessel A-1 because the juncture of the center lines was used as the theoretical origin.

Test Results in the Plastic Range

Curves representing the technique for analyzing the residual stresses after plastic strains pertaining to vessel A-1 are shown in Figs. 4, 5, 6 and 7. Two of these curves, Figs. 5 and 7, are modified stress-strain curves for vessel A-1 and were obtained as described in a previous publication (*op. cit.*) The simple stress-strain curves were modified to account for the combined stress condition existing in the vessels. This was done by using

Fig. 3 General view of forty-eight inch diameter vessel

the distortion energy theory to obtain the proportional limits for the modified stress-strain curves and the maximum shear theory to obtain the horizontal portions, which occurred at general yielding, for the combined stress conditions. The load that caused general yielding of the cylindrical portion of the vessel was used to obtain the general yield stress for each vessel. At the present time, use of the distortion energy theory and the maximum shear theory to obtain the stress at the proportional limit and general yielding point, respectively, is considered a reliable engineering practice. The above procedure, although giving trends to quantitative results within engineering accuracy, is open to further work on the "properties of materials."

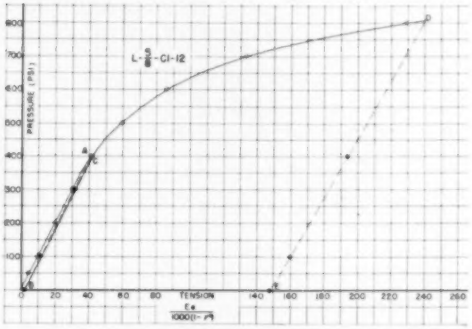


Fig. 4 Measured strains, Vessel A-1, gage L-5/8-CI-12

The dotted lines of the modified stress-strain curves for vessel A-1 (Figs. 5 and 7) show the magnitude and reversal of the residual stresses on the cone adjacent to the juncture. This reversal occurred on the inside of the cone 5/8 in. from the origin (juncture) and the residual stress varied from the yield in tension to the yield in compression. An even greater residual compressive strain might be measured if it were possible to place gages closer to the origin without their being affected by the weld. It is evident that the residual stresses obtained are in the transition zone between the proportional limit and the general yield point. Thus, they are not exact because only the end points of this transition zone have been obtained by the accepted engineering theories. In order that the transition zone may be more completely understood, it will be necessary to conduct extensive tests directed toward this particular end.

The force diagram shown in Fig. 8 indicates that the force balance at the cone side of the juncture is consistent with statics. This consistency proves that the electrical resistance strain gages give satisfactory results during the early stages of plastic yielding. (op. cit.)

A comparison of the above results for vessels A-1 and A-2 indicates that the plastic strains were much greater at the sharp junction of vessel A-1 resulting in a reversal of stress from yield in tension to yield in compression. Thus it follows that a fatigue type of failure is indicated for a repeated load when the cone to shell juncture is

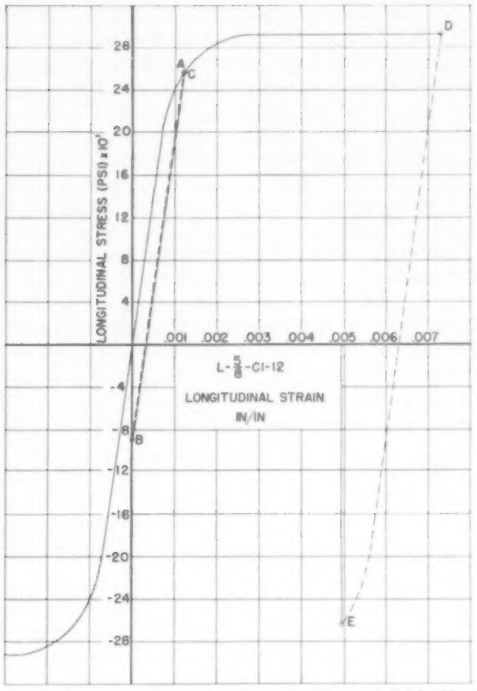


Fig. 5 Stress-strain curves, Vessel A-1, gage L-3/4-CI-12

relatively sharp. The results obtained for vessel A-2 indicate that a substantial knuckle radius will tend to correct and eliminate this possibility.

CONCLUSIONS

48-In. Diameter Vessels

1. Because the material in the welds had a higher yield than the material in the plates, uniform general yielding was prevented on the cylindrical section when

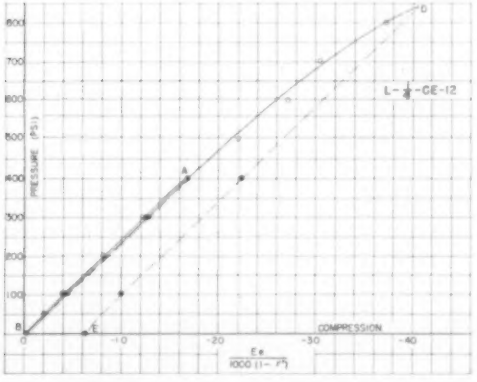


Fig. 6 Measured strains, Vessel A-1, gage L-1/4-CE-12

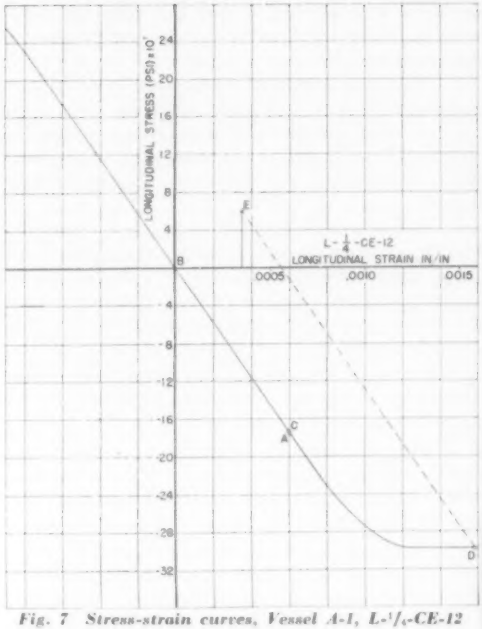


Fig. 7 Stress-strain curves, Vessel A-1, L-1/4-CE-12

the pressure was sufficient to give a circumferential stress equal to the yield strength for the cylindrical plates.

2. The maximum plastic strain in vessel A-1 occurred adjacent to the sharp cone-shell juncture and not at its theoretical center. Since the theoretical center fell within a circumferential weld, it follows that

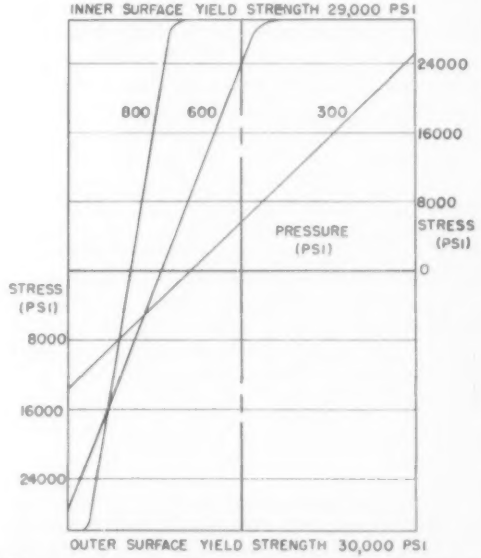


Fig. 8 Balance of forces from measured stresses at knuckle, Vessel A-1, L-3/4-CI-12, L-1/4-CE-12
Inner surface on the right Outer surface on the left

welding at the cone-shell juncture minimizes the plastic straining ordinarily expected, but gives rise to higher stresses in the weld.

3. The maximum plastic strain in vessel A-2 occurred at the cone side of the torus section.

4. Internal pressure vessels of a comparable D/t ratio may be supported in a horizontal position without effectively altering the directions of the principal stresses.

5. The allowable working pressure for vessel A-2 was 335 psi. based on an ultimate stress of 55,000 psi. and a safety factor of 4. Since local yielding occurred at nearly 600 psi. (about one and three quarters times A.S.M.E. allowable pressure) stresses beyond the yield strength may be expected in the longitudinal direction at a juncture where the cone and knuckle thicknesses are slightly greater than that of the side walls before a test pressure of twice the allowable is reached.

6. When pressures equal to two and one-third times the A.S.M.E. allowable working pressure are applied to vessels with conical or toriconical heads, general plastic yielding occurs in the cylinder walls.

Retest of 8-In. Diameter Model No. 2

Test Preparation. Eight-inch diameter Model No. 2 had been stored in the laboratory under heat lamps for a period of approximately seven months. The heat lamps were kept on continually in order that humidity conditions would be reduced, thus making it possible to utilize the same strain gages that had been used in the first test on this model. To prepare this vessel for test, it was merely a question of determining whether or not the electrical resistance strain gages were intact. The resistance to ground was checked and found to be in ex-

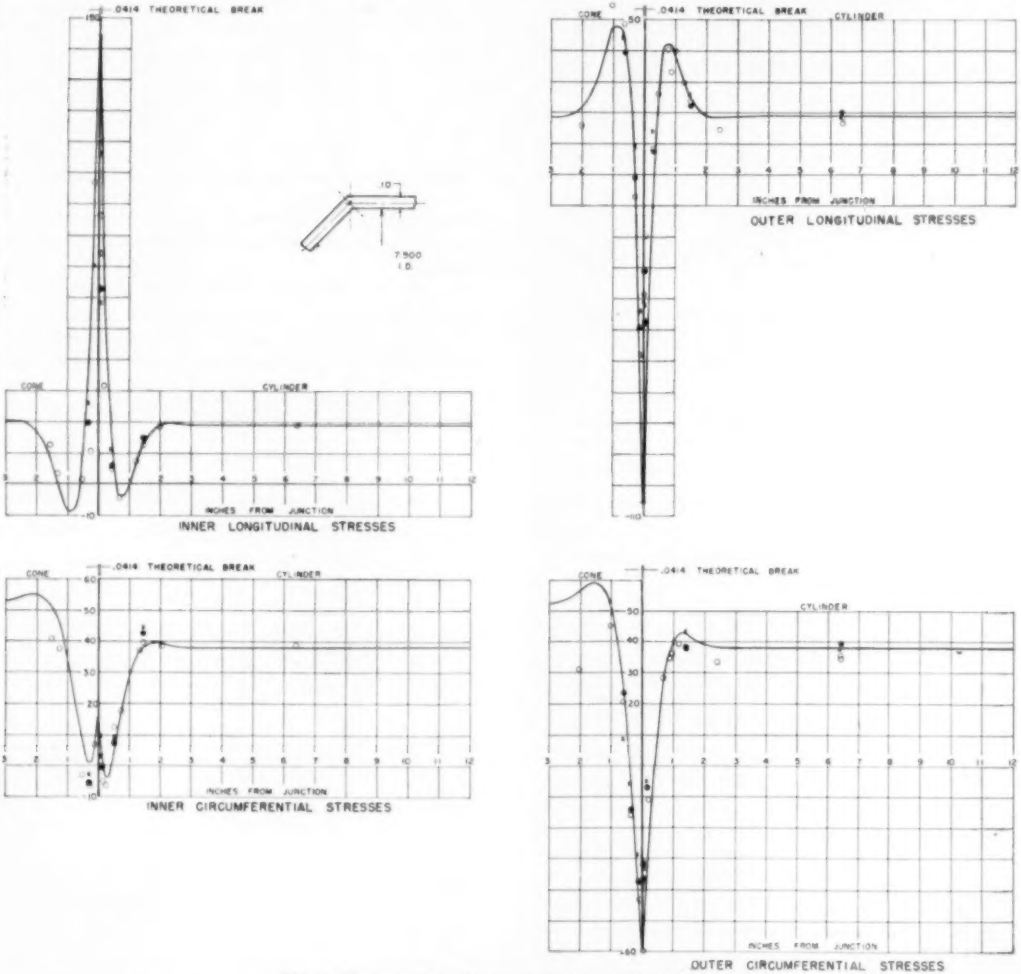


Fig. 9 Computed and measured stresses, Model No. 2

Stress based on an internal pressure, P, of 1 psi. Experimental stress measurement: ○ indicates first test; ● indicates second test

cess of 100 megohms for both internal and external strain gages. The gages were then wired into the 100-channel switching unit as before and the piping system was completed to a small hand pump.

Test Procedure. The testing procedure for this model was, generally speaking, much the same as that employed for all other tests. The pressure was raised in gradual stages to approximately half the pressure calculated for generalized yielding. Released through gradual stages, the pressure was again raised to a point close to generalized yielding and, for each increment of pressure, strain gage readings were taken. By following this technique, it was possible to obtain strain values in the plastic region for all strain gages.

Results. The results obtained from testing Model No. 2 were plotted in the form of pressure strain curves; these curves in turn yielded stress values which could be plotted against computed results. The measured vs. computed stress curves may be found in Fig. 9 where the solid circles indicate the stresses measured during the second test. This retest on Model No. 2 was made in order that the effect of time after plastic strain might be determined.

EVALUATION OF DESIGN DIVISION RESEARCH PROGRAM AT PURDUE UNIVERSITY†

While it is extremely difficult for anyone directly involved in a research project to evaluate the results of that research while the work is still in progress, it is possible to indicate a few of the accomplishments and conclusions resulting from the investigations:

1. When the P.V.R.C. study of the head problem was initiated in 1945–46, each of the several approximate analyses then in use for straight or curved shells had a considerable following of its own. A literature survey and comparative studies showed that, except in nomenclature, these theories were essentially identical, being approximations of the theoretically more accurate Love-Meissner analysis. The use of the latter was proposed at that time by W. R. Burrows of the Standard Oil Co. (Indiana) to serve as an index of the degree of accuracy of shorter methods, and as a basis for the preparation of curves indicating the stresses and deformations in well-known head shapes in a range of proportions.

2. The literature survey, previously referred to, revealed little reliable published data giving actual measured stresses on the inside as well as the outside of a pressure vessel.

3. The tests of the first model pressure vessel confirmed that SR-4 gages can be insulated against water for relatively low pressure (less than 300 psi.), but no suitable method has thus far been found for sustained insulating of the gages at higher water pressures.

4. By the use of a dielectric transformer oil as a medium for applying pressure to the test vessels, pressures up to 2200 psi. were reached without destroying the bond between the gages and the vessels, or short circuiting the electrical circuits of the gages. In calibration tests on gages, pressures up to 6000 psi. were used successfully with the dielectric transformer oil as the pressure-producing medium.

5. Preparation of the test models revealed, in general, that these small vessels can be machined to very close tolerances in the head-to-shell juncture region, either with a definite knuckle radius or without a knuckle. The results of tests on these vessels indicated that when the radius was very small (case of conical head with no knuckle) the measured elastic stresses were in close agreement with those computed either by the Love-Meissner analysis (W. R. Burrows) or by approximations of the Love-Meissner analysis of the Boardman or Wetterstrom type (E. Wetterstrom). But, for vessels with an appreciable knuckle radius, (toriconical heads), the agreement with approximate methods in the vicinity of the knuckle was not quite so good, because no adequate approximations of the Love-Meissner analysis for curved shells had been developed.

6. The peak stresses measured in the knuckle region of toriconical heads were somewhat smaller than those predicted by the approximate theory for conical heads. When no knuckle radius was present, the measured stresses were very close to those predicted by the approximate theory for conical heads.

7. A more complete analysis, which accounts for variation of stress at the head-cylinder juncture region, has been suggested by W. R. Burrows of the Standard Oil Co. (Indiana). The test results at the juncture indicate a closer agreement with the results of this more complete analysis than with the results of the approximate theory for conical heads which makes no adjustment at the juncture. Otherwise this approximate theory appears to be quite satisfactory.

8. The test results indicated that plastic action in the knuckle region of the vessels studied started at pressures near the allowable working pressures determined by the A.S.M.E. Code for Unfired Pressure Vessels.

9. Retests of the 8-in. diameter model pressure vessel No. 2, seven months after considerable plastic deformation had been produced by the initial test, indicated that the elastic stresses in the vicinity of the cone-cylinder juncture were of the same general character and distribution found in the test of the original annealed and unstressed vessel.

10. The tests on the 48-in. diameter vessels revealed that the general trend of stresses in these larger vessels was in close agreement with those in the models and those indicated by the approximate theory of conical heads where applicable. Thus the latter appears to be sufficiently exact for engineering use. This is not true, however, for existing approximate theories for toriconical heads.

11. Fabricating irregularities, after surface grinding to the extent encountered in these vessels, caused only

† The following information was first presented and reviewed at the May 10, 1949, meeting of the Design Division of P.V.R.C.

slightly greater stresses than occurred in comparable places where no irregularities existed.

12. The vessel tests, as well as the material property tests, reaffirmed the general understanding that the yield point of the material in the weld metal is higher than at points in the vessels some distance from the weld.

13. The fact that the welds had a higher yield strength than the base metal indicated that if welds are placed at the sharp cone-cylinder junctures, there will be less localized yielding and higher absolute values of the tensile stresses than there would be if there were no welds at the junctures. For static loads, the effect of the welds may be beneficial, but for shock or repeated loads the margin of safety would be expected to be less than if the welds were placed a short distance from the junctures, a difficult fabrication procedure.

ACKNOWLEDGMENTS

The authors are indebted to W. R. Burrows, Chairman, and other committee members of the Design Division of the Pressure Vessel Research Committee, which is sponsored by the Welding Research Council of the Engineering Foundation, for aid rendered throughout the course of this investigation. The authors are also indebted to Dr. A. A. Potter, Dean of Engineering, and Professor R. B. Wiley, Head of Civil Engineering and Engineering Mechanics, for administrative assistance in conducting the research at Purdue University.

Determination of the Temperature of Sprayed Metal Particles

(*Continued from page 322-s*)

released by cooling from the melting point to T_s. The remainder of the 6540 calories, or 2390 calories, comes from the heat of solidification of the molten copper. As the heat of solidification for copper is 3120 cal./mol., it is assumed that only a portion of the particles is in the molten state. Since only 104 calories or 1.7% of the total heat can be attributed to the change in kinetic energy, the particles must be molten even before impact occurs. The energy of impact is insignificant as far as the determination of the spraying temperature is concerned.

If the impinging particles are at their melting points, as the data indicate, a plausible explanation of the structure of sprayed deposits is possible. At its melting point the metal is highly susceptible to oxidation so that a skin of oxide forms on the particle in flight. At impact the mushy, plastic particle with its coating of oxide offers little resistance to deformation. As a result, the particle is flattened into a thin platelike section which assumes the contour of the surface beneath it. The oxidized surface is trapped between the particles, giving the total structure a stratified appearance when viewed at high magnification. Molybdenum exhibits this structure very markedly, as may be seen in Fig. 1 of previous article.[9] This deposit has a density of 85% of theoretically dense molybdenum and an oxygen content of 3.5%.[9]

The appearance of occasional spherical particles is ascribed to the fact that although the average particle is at the melting point, some particles may be above the melting point and others, below. Those that are cooled below their melting point before impact resist deformation and retain their spherical or teardrop shape.

CONCLUSIONS

(*a*) A method has been developed for determining the heat content and temperature of sprayed metal particles at the moment of impact.

(*b*) The results obtained by this method indicate that sprayed particles of low- and high-melting-point metals are at their melting points at impingement.

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Underbead Cracking of Welds Cathodically Charged with Hydrogen

Studies of retained austenite and the hydrogen concentration in the weld and heat-affected zone on underbead cracking. Individually they do not produce underbead cracking. Together they do

by M. W. Mallett and P. J. Rieppel

INTRODUCTION

THE problem of underbead cracking has been the subject of extensive study by the present authors as well as numerous other investigators (1–5). Underbead cracks occur in the base metal close to, and generally parallel to, the fusion line of welds made in alloy steels of moderate and high hardenability and in some carbon-manganese steels. The hydrogen theory of underbead cracking (3, 4) is now commonly accepted, though the retention of austenite at room temperature and its transformation are necessary to an understanding of its mechanism (2). Occasionally such cracking is still discussed by some as a function of hardenability alone.

The mechanism, as expressed by Hoyt, Sims and Banta (2) is that underbead cracks are caused by the retention of austenite in the heat-affected zone, and the precipitation of hydrogen from the residual austenite during its transformation to martensite at low temperature. Toe cracking very often is present with underbead cracking, and is attributed to the same mechanism. There is much indirect evidence that hydrogen absorbed by the weld metal and heat-affected zone from the arc atmosphere is the chief cause of underbead cracking and toe cracking (3, 4, 6). Mallett (7) has shown that the hydrogen content of the weld metal at the time of solidification can be calculated rather accurately from the partial pressure of hydrogen in the arc atmosphere and Sieverts' solubility data for 2820° F. Likewise, work at Battelle Institute (1) has shown a good correlation between the hydrogen content of the arc atmosphere and the extent of underbead cracking in a crack-sensitive steel. Flanigan (8), in his study of the ductility of arc welds in mild steel, emphasized the necessity of isolating the role of hydrogen.

Cracking of weldments is obviously produced by stress, of which there appear to be three sources. These are: (*a*) shrinkage of weld and heat-affected zone; (*b*) the austenite-martensite transformation; and (*c*) hydrogen. Once the weldment has reached room temperature, the shrinkage stresses remain more or less constant. The present investigation was, therefore, concerned with studying separately the effects of the two variable factors; that is, the transformation of retained austenite and the hydrogen concentration in the weld and heat-affected zone. The experiments were designed so that the austenite-martensite transformation could progress to varying degrees of completion before hydrogen stresses were introduced. This was achieved by making welds with low-hydrogen welding electrodes; annealing and stress relieving the welds as desired; and then saturating the test plates with hydrogen by cathodic charging.

EXPERIMENTAL PROCEDURE

In general, the test procedure consisted of making weld beads on a crack-sensitive steel with low-hydrogen electrodes, which of themselves do not produce underbead cracking. After storage at room temperature for various periods of time, or after suitable heat treatment or low-temperature treatment to promote the austenite-martensite transformation, the welded specimens were cathodically saturated with hydrogen, and then sectioned and Magnaflux inspected for cracks.

The test specimens were 1 in. thick, 3 in. wide and 6 in. long. Most of the tests were made on heat-treated manganese-molybdenum steel (C 0.25–0.28%, Mn 1.50%, Mo 0.50–0.60%). A few tests were made on 18–8 stainless steel plate, ship plate steel (C 0.18%, Mn 1.18%, Si 0.28%) "as-received," and on homogenized samples of manganese-molybdenum steel and ship plate.

A weld bead 4 in. long was deposited on the longitudinal center line of the broad face of the specimen, starting 1 in. from one end, as shown in Fig. 1 (*A*). The specimen was at room temperature when welding was started. The majority of the welds were made with a $^{3}/_{16}$-in. diameter low-hydrogen type ferritic electrode.

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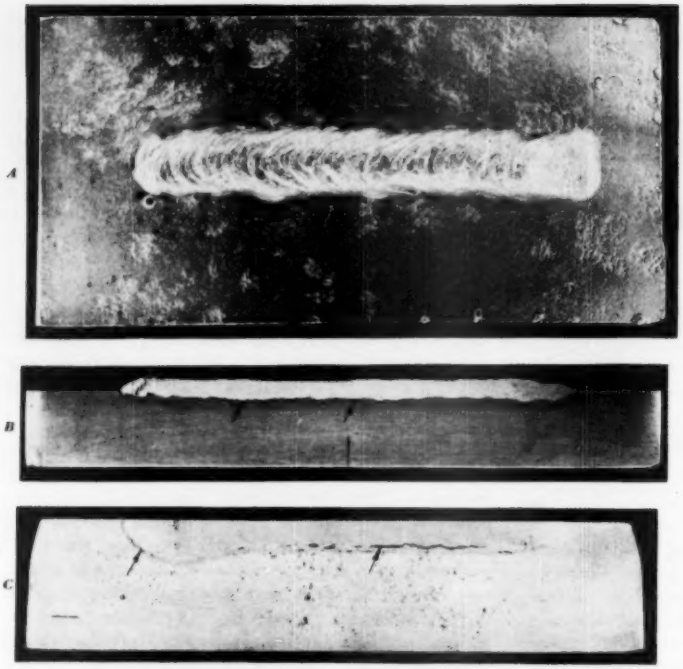

Fig. 1 Typical example of welded specimens and sections showing underbead cracks and toe cracks

(A) Typical welded specimen; (B) section of specimen showing underbead cracks; (C) sectioned specimen prepared to show toe cracks

A few welds were made with $^3/_{16}$-in. 18–8 stainless electrodes. The welding beads were made manually at about 7 in. per minute, using 180 to 200 amp. d. c. reverse polarity.

All except the various control specimens were cathodically charged with hydrogen. Specimens to be charged immediately after welding were allowed to cool in air for 30 to 60 min. to approximately room temperature before placing in the charging cell. Other specimens were held at about 70° F. for periods of 24 hr. to 6 days, or were subjected to heat treatment or liquid-air treatment to hasten the austenite-martensite transformation before charging.

From one to three specimens were charged simultaneously. The specimens formed the cathode, while lead-plate anodes were located on both sides of the specimen to give uniform current distribution. The electrolyte consisted of about 10 gal. of 10% H_2SO_4 containing a 1-gm. addition of As_2O_3 to act as a promoter and increase the hydrogen absorption of the specimen. Small amounts of impurities such as arsenic are reported (9) to increase hydrogen absorption one hundred times. The current density was 0.3 amp. per square inch, and the operating bath temperature 110° F. Charging was continued for 6 hr. to insure saturation.

Upon removal from the charging bath, the specimens were washed in cold water to remove the acid, and then sectioned longitudinally through the center of the bead, polished on a belt grinder, and Magnaflux inspected for underbead cracks, as shown in Fig. 1 (*B*). Permanent records of the crack indications were made by lifting the tracing with transparent scotch tape and fastening the tape to record cards. The percentage of cracking was calculated from the total length of cracking per length of bead. Later, the remaining part of the weld bead was ground flush with the surface of the plate, and the specimen was Magnaflux inspected for toe cracks, as shown in Fig. 1 (*C*).

EXPERIMENTAL RESULTS

The results of the tests on the effect of cathodic charging with hydrogen on the underbead cracking of weldments are given in Table 1. The manganese-molybdenum steel plate was chosen because it was quite crack sensitive. The electrode used was picked because of its low hydrogen, noncracking characteristics. Weldments with this combination of electrode and base plate showed no underbead cracking when sectioned immediately after welding or after storage for 60 days, as shown in Fig. 2 (*A*). If cathodically charged immediately after welding, 100% underbead cracking was produced, Fig. 2 (*B*). Homogenizing the armor plate (heated at 2200° F. for 4 hr.) before welding had no significant effect on its cracking tendencies. When welded specimens were stored for 24 hr. before charging, only 50% cracking resulted. Six-days' storage before charging entirely eliminated the tendency toward underbead cracking. However, toe cracking remained close to maximum. The degree in which underbead cracking diminishes with time appears to follow changes which occur in the heat-affected zone of the weld. According to theory, one such change is the gradual transformation of retained austenite to martensite. Therefore, these data substantiate the theory that stresses imposed by hydrogen are added to those accompanying the austenite-martensite transformation, and underbead cracking results. If the hydrogen is not present in the retained austenite, no underbead cracking occurs from the austenite-martensite transformation alone. Likewise, if the austenite is transformed before hydrogen is added, no underbead cracking occurs.

When welded specimens were subjected to treatments designed to transform austenite, such as heating at 1000° F. for 1 hr. or by supercooling in liquid air 1 hr., no underbead cracking was produced by cathodic charg-

Table 1—Results of Tests on Underbead Cracking of Welds Electrolytically Charged with Hydrogen

Test No.	Electrode used	Base plate	Treatment after welding	Charged with hydrogen*	Cracking in 3 specimens†, % Underbead	Toe
3	LH‡	Mn-Mo steel§	None	None	None	None
1	LH‡	Mn-Mo steel§	None	Immediately after welding	100	100
2	LH‡	Mn-Mo steel§	None	24 hr. after welding	50	80
5	LH‡	Mn-Mo steel§	None	6 days after welding	None	90
4	LH‡	Mn-Mo steel§	Heated at 1000° F. for 1 hr.	About 3 hr. after heat treatment	None	None
6	LH‡	Mn-Mo steel§	Cooled in liquid air 1 hr., warmed to room temperature	About 6 hr. after cooling in liquid air	None	90
7	LH‡	Mn-Mo steel§	Cooled in liquid air 1 hr., warmed to room temperature	None	None	None
11	LH‡	Mn-Mo steel§	None	Weld only, immediately after welding#	30	40
12	LH‡	Mn-Mo steel§	(Plate charged before welding)	None	5	30
14	LH‡	Mn-Mo steel (homogenized)	None	Immediately after welding	100	100
8	18–8¶	Mn-Mo steel§	None	Immediately after welding	None	20
10	18–8¶	Mn-Mo steel§	None	Weld only, immediately after welding#	None	None
13	18–8¶	18–8 steel	None	Immediately after welding	None	None
9	LH‡	Ship-plate steel	None	Immediately after welding	None	None
15	LH‡	Ship-plate steel (homogenized)	None	Immediately after welding	None	None

* Charged electrolytically in H_2SO_4 for 6 hr.
† Specimens were 6 x 3 in. with a weld 4 in. long. The electrodes were $^3/_{16}$ in. and the weld was made manually at 6 in. per minute using 180 to 200 amp. reverse polarity.
‡ Low-hydrogen ferritic electrodes.
§ The Mn-Mo steel was heat treated. See Experimental Procedure paragraph 2.
# Charged weld only, the remainder of the weld was coated.
¶ Steel used was 18–8 stainless steel.

ing. However, toe cracking did develop when specimens pretreated in liquid air were charged. This does not mean that underbead cracking and toe cracking in normal welds are produced by independent mechanisms. Experience has shown that toe cracking is not produced separately except under unusual conditions such as those present in tests in which the retained austenite was transformed after welding, and then the test piece heavily charged with hydrogen. Although, in this case, toe cracking might be attributed to stress corrosion, it should be noted that much stress-corrosion cracking is probably the result of hydrogen embrittlement brought about by corrosion.

One group of weldment specimens was coated, except for the weld bead, with gray Stonite (Stoner-Mudge, Inc.), a vinyl-type plastic, before charging so that hydrogen could be absorbed by the weld only. These specimens were coated and charged immediately after welding. Apparently, because of the reduced amount of hydrogen entering the specimens, only 30% underbead cracking occurred.

When the manganese-molybdenum base plate was cathodically charged with hydrogen immediately before welding, it retained enough hydrogen to produce 5% underbead cracking and 30% toe cracking without further charging.

Weldments made with 18–8 stainless steel electrodes on armor plate did not produce underbead cracks when charged with hydrogen immediately after welding.

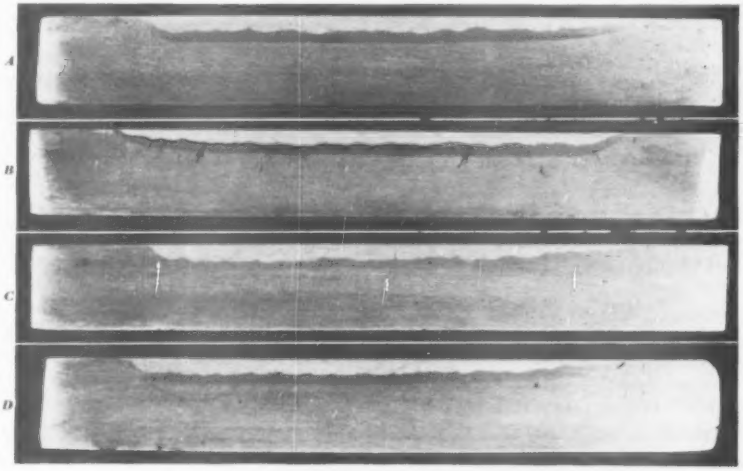

Fig. 2 Series of weld samples showing influence of hydrogen, stress relief, and austenite-martensite transformation on underbead cracking

All welds were made with low-hydrogen type electrodes. The weld beads were ground flush for later inspection for toe cracks.
(A) Section of specimen "as-welded"; (B) section of specimen charged with hydrogen immediately after welding showing underbead cracks; (C) section of specimen stress relieved after welding and then charged with hydrogen; (D) section of specimen treated as follows: (a) cooled in air; (b) quenched in liquid nitrogen to transform retained austenite; (c) warmed to room temperature; (d) charged with hydrogen

However, about 20% toe cracking was found in one series of specimens, as shown in Table 1. The reasons why no underbead cracking resulted from these tests is not understood, since the heat-affected zones of the welds made with 18–8 electrodes should be very similar to those of welds made with ferritic electrodes, and both types of specimens received the same hydrogen-charging treatment. Possibly the weld stresses, resulting from the 18–8 welds, were lower than those resulting from the higher strength ferritic welds, and, therefore, no cracking resulted.

In connection with these specimens, an interesting phenomenon was noticed. Apparent indications of cracking were given by the Magnaflux test. However, closer inspection revealed that no cracks were present, but thin layers of entrained ferritic base metal mixed with nonmagnetic austenitic weld metal produced false indications of cracking within the weld bead.

No cracking was produced by immediate cathodic charging of weldments of 18–8 electrodes on 18–8 base plates, or low-hydrogen ferritic electrodes on ship-plate steel, "as received," or homogenized.

DISCUSSION OF RESULTS

No underbead cracking or toe cracking occurred in any of the control (uncharged) specimens where the hydrogen content of the metal was known to be low or essentially nil. Moreover, no cracks resulted from hydrogen-charged specimens in which there were large amounts of austenite which did not transform or very small amounts of transient austenite when charged; for example, the 18–8 electrode on 18–8 base plate, the samples chilled in liquid air, and welds made on ship plate which normally shows only a little underbead cracking. There appears to be a direct correlation between the amount of austenite available for transformation after cathodic charging with hydrogen, and the amount of underbead cracking. Also, previous work has shown that when welding conditions which produce retained austenite were constant, the degree of cracking was proportional to the concentration of hydrogen in the arc atmosphere or to the concentration of hydrogen in the deposited metal. Thus, it appears from separating the component causes of underbead cracking that they do no harm individually, but when present concurrently, they cause cracking to the degree in which each factor is present.

Toe cracking occurred in specimens which did not contain underbead cracks; for example, those cooled in liquid air before charging and those stored in the as-welded condition several days before charging. The specimens stress relieved at 1000° F. did not contain toe cracks or underbead cracks. It appears then, that the toe cracking was a result of weld stresses at the toe of the weld bead and hydrogen embrittlement from cathodic charging with hydrogen or possibly from stress corrosion. Under ordinary welding conditions, stresses induced by the austenite-martensite transformation are known to contribute to toe cracking. On the other hand, from this work, weld stresses plus hydrogen do not appear to be enough to produce underbead cracking in the absence of transforming austenite. Relatedly, toe cracks may result from pickling weldments in acid even after they have been stored long enough for the retained austenite to transform, and there is no longer danger of producing underbead cracking.

PROPOSED THEORY OF UNDERBEAD CRACKING

A theory of underbead cracking is proposed on the basis of results obtained in the present investigation. The theory is based on the following premises:

1. A high-hydrogen concentration in stable austenite does not cause cracking.
2. Underbead cracking is not dependent on the total amount of retained austenite, but rather on the amount of it which transforms several hours to several days after the weld reaches constant temperature (below about 200° F.) Moreover, hydrogen must be present in sufficient concentration or cracking does not occur.
3. Underbead cracking appears to occur only while retained austenite is undergoing the austenite-martensite transformation.
4. Underbead cracking does not occur when hydrogen is added to the resulting martensite, after transformation, even though such subsequent addition may embrittle the martensite.
5. Cracking occurs only during kinetic stressing of local areas.

Briefly, underbead cracking is brought about through: (*a*) the embrittling effects of hydrogen; and (*b*) the dynamic stresses of the austenite-martensite transformation superimposed on shrinkage stresses.

The Role of Hydrogen

A weldment made with a hydrogen generating electrode, such as the E6010, may contain excessive hydrogen. When cooled to room temperature, it will evolve hydrogen for some hours after the welding is completed. Thus, the weld and heat-affected zone are saturated or supersaturated with hydrogen almost immediately upon welding, and for several hours thereafter. Hydrogen also fills pre-existing submicroscopic rifts and gross voids without causing cracking.

During the austenite-martensite transformation, described below, hydrogen, which has a lesser solubility in martensite and ferrite than in austenite, is rejected from the transformed austenite, and, due to local increase in hydrogen pressure, must diffuse in part: (*a*) away from the austenite and become harmless, and (*b*) into the remaining retained austenite and increase its hydrogen content. At the instant of transformation, the metal lattice is probably in a state of greatest disorder. Part of the rejected hydrogen precipitates as molecular hydrogen into microvoids in the structure building up high aerostatic pressures. This tends to

stabilize these imperfections and distortions in addition to embrittling the martensitic transformation product by interstitial solution of hydrogen in the atomic lattice.

The work of Herres (10) indicates that fully hardened martensite is ductile when low in hydrogen, but becomes completely brittle when charged cathodically with hydrogen. Likewise, the martensite is brittle if the steel is charged by heating in hydrogen for 1 hr. at 1600° F. before quenching.

The mechanism by which fracture of the embrittled martensite of the weld region takes place is discussed in the following section.

The Role of the Austenite-Martensite Transformation

The concept of room-temperature isothermal transformation of retained austenite has been rejected by some and accepted with reluctance by others; this is in spite of the work of J. A. Mathews (11) in the 1920's and of others, and of the evidence on S-curves. However, the work of Averbach and Cohen (12) has shown not only that the phenomenon exists, but that retained austenite can be measured quantitatively. Recently, Averbach, Castleman and Cohen (13) have stated that "where low-carbon steel is effectively austenitized at high temperatures and then drastically quenched by the adjacent mass of metal, it now appears that austenite may be retained in measurable amounts. There is also good reason to believe that such austenite undergoes isothermal transformation at room temperature, as has been observed for high-carbon steels, and thus may have a bearing on the phenomenon of underbead cracking." The role of this transformation became clear in the original work (2), and its reality seems no longer open to doubt.

As to the nature of the austenite-martensite transformation, Hultgren (14) states that, "it is generally held that martensite is body-centered iron with carbon in enforced solution formed by some shearing movement, without diffusion, and, in consequence, having the same composition as the mother austenite." Now, if the austenite is saturated with hydrogen it also will be trapped interstitially at the instant of transformation, embrittling the resulting martensite. Some of the hydrogen will subsequently diffuse in the manner described in the previous section.

Volume changes accompanying the austenite-martensite transformation apparently create stresses* which are sufficient to cause cracking of the hydrogen-embrittled martensite formed by the first portion of the austenite to transform. When the same transformation takes place in material low in hydrogen, the martensite is ductile enough to permit the stresses to be relieved by elastic deformation. The above considerations are enough to account for cracking or noncracking in the underbead region.

Discussion of Theory

Parts of the above concept of the behavior of hydrogen have been described elsewhere. The idea of hydrogen stabilizing transitory inhomogeneties by precipitation into dislocations in the lattice is new. The present theory is an attempt to join the various fragmentary hypotheses into a logical and understandable unit. The experimental work has provided good substantiating results for previous theories, and filled in important gaps in the data.

CONCLUSIONS

Conclusions based on this investigation, certain previous work and a proposed theory are as follows:

1. Underbead cracking does not occur in weldments free of material undergoing the austenite-martensite transformation.

2. A direct correlation exists between the amount of austenite retained at the time of cathodic charging with hydrogen, and the amount of underbead cracking.

3. Previous work (1) has shown that when welding conditions which produce retained austenite are constant, the amount of underbead cracking is proportional to the concentration of hydrogen in the weldment.

4. The effects of the variable component causes of underbead cracking: (*a*) retained austenite; and (*b*) concentration of absorbed hydrogen have been isolated. Individually they do not produce underbead cracking. Together, they do. Because of controlled welding conditions, weld shrinkage stresses were approximately equal for compared specimens.

* It appears that in the austenite-martensite transformation, groups of atoms assume a new lattice structure with great rapidity. This suggests that, perhaps, the accompanying stresses (in submicroscopic areas) are dynamic in nature and consequently, effectively greater than calculated static stresses due to the transformation. However, there seem to be no data presently available on the increase in stress to be expected from this source.

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Mechanical and Material Variables Affecting Correlation

Mechanical and material variables which affect correlation of performance between small-scale specimens and complex structures. The relative performance of V-notch Charpy and nick-bend specimens as compared with the performance of welded 9-ton box girders. A complex interplay of mechanical and material variables precludes correlation except in a very approximate sense

by Carl E. Hartbower and William S. Pellini

INTRODUCTION

Statement of the Problem

EXPERIENCE has shown that for certain applications tensile properties obtained with the conventional tensile specimen are not sufficient to evaluate a steel as to its behavior in service. Numerous investigations have been aimed at developing tests to supplement the tensile specimen for quantitative prediction of performance in monolithic structures. Prediction of performance is based on the premise that a rigorous correlation exists between the behavior of small-scale laboratory tests and that of full-scale structures. In general there are two approaches to the problem of establishing such a correlation. One is by a systematic investigation of small-scale tests run concurrently with tests on large structural members taken to represent elements of a full-scale structure, and the second is by post-mortem, small-scale testing of material taken from actual casualty structures.

Bending tests of four 9-ton welded box girders of identical design and from a single heat* of fully killed steel were conducted at the National Bureau of Standards over a range of temperatures. The Naval Research Laboratory was requested to attempt a correlation between the performance of the box girder and various small-scale laboratory specimens. The small-scale tests were to include standard and high-constraint nick-bend specimens tested in slow-bend and V-notch Charpy specimens tested in slow bend and in impact.

It should be noted that this investigation was primarily concerned with only one steel and only one design of prototype structure. With such a limitation it was outside the scope of the program to seek a general correlation between laboratory specimens and service performance of structures. The course taken was therefore limited to a critical investigation of the variables affecting the laboratory phase of correlation. It will be shown that rigorous representation of performance by relatively simple laboratory specimens is itself an extremely complicated problem which sould logically continue to defy a nonstatistical approach.

Related Investigations

The investigation at the Naval Research Laboratory was made in conjunction with a research program at the National Bureau of Standards under the auspices of the structural Steel Research Committee of the Welding Research Council.[1] The main objective of the Bureau of Standards program was to determine the effect of a combination of high-residual stresses and severe constraint upon the capacity of a structural member to resist rupture under external load at service temperatures likely to be encountered by a structure in the open, such as a bridge or a ship. The prototype structure chosen for test consisted of a box section (Fig. 1) which was deemed to simulate critical design features of ship sections and other large welded structures. The details of the box girder intentionally did not conform to good design practice since it was intended to intensify the unfavorable restraint and locked-in stress features of welded monolithic structures. To provide additional constraint and to introduce greater shrinkage effects, the sizes of the welds, particularly the fillets and the reinforcement, were made considerably larger than necessary. Moreover, a welding sequence was chosen to set up as severe a condition of restraint against weld shrinkage as possible without actually introducing weld cracking.[2]

Radiographic examination of the weld joints in the vicinity of mid-span of Girders 1 and 2 (see Table 1 for notation) was made by the Naval Gun Factory to establish the general quality of the welds. Forty-four locations in Girder 1 and 48 locations in Girder 2 were examined.

Table 1—N.B.S. Box-Girder Tests

Box girder	*Temp. of test, ° F.*	*Maximum load, Kips.*	*Moment, in. Kips.**	*Modulus of rupture, Kips./in.²**	*Maximum deflection, in.*	*Fracture*
BG1	+80	1397†	71,200	75.6	8.00	Cleavage
BG2	−40	1165†	59,400	63.1	2.45	Cleavage
BG4	0	1476†	75,300	79.9	8.83§	Cleavage
BG5	+40	1670‡	85,200	90.4	16.16	None
BG3	+80	1685‡	85,900	91.2	18.06	None

* Computed from the load at failure or from the maximum load that was reached.
† Load at failure.
‡ Maximum load that was reached. No failure.
§ At 1400 kips load and 6 in. deflection a crack was detected in the side plates. Further loading to 1476 kips resulted in sudden rupture.

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The opinions contained herein are those of the authors and are not necessarily to be construed as official or reflecting the views of the Navy Department or the Nava Service at large.

* Except one pilot specimen constructed of semikilled steel.

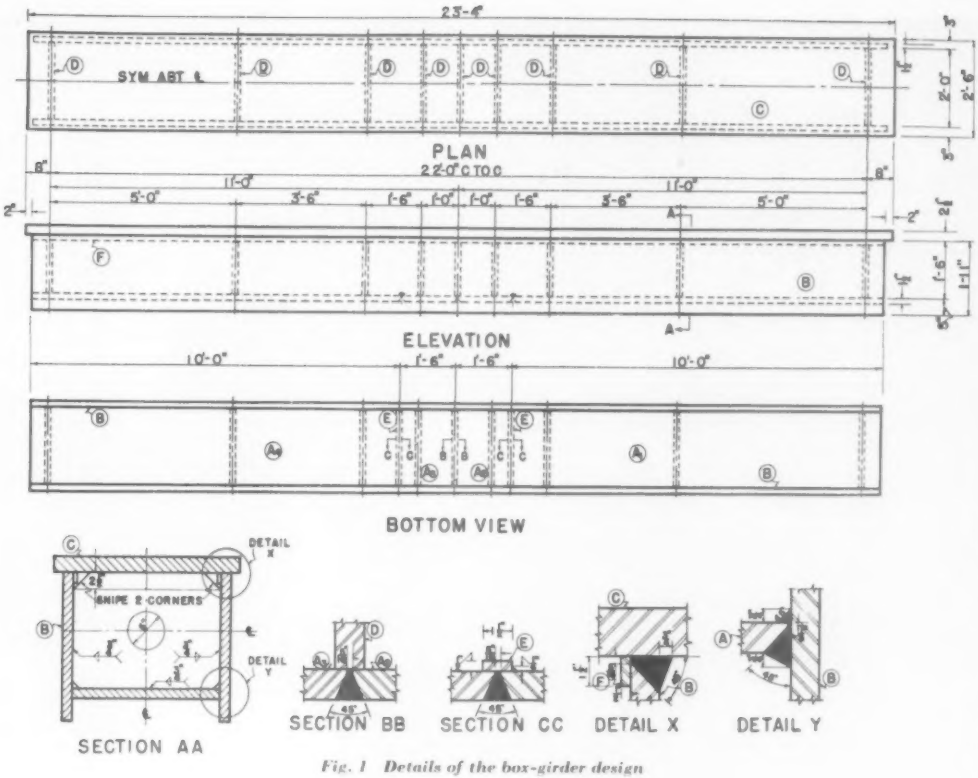


Fig. 1 Details of the box-girder design

Table 2—Chemical Analysis of the Box Girder Plates

BOX GIRDER NO.	PLATE	C	Mn	Si	S	P	Cu	Ni	Cr	Mo	Al	N	Ti	V
								(a)	(a)	(a)	(a)		(a)	(a)
BG1	A	0.24	0.53	0.05	0.020	0.015	0.02	0.01	<0.01	<0.01	0.02	0.011		
Semikilled	B1	0.24	0.57	0.06	0.030	0.015	0.02	0.01	<0.01	<0.01	<0.01	0.013		
	B2	0.25	0.56	0.05	0.049	0.015	0.02	0.01	<0.01	<0.01	<0.01	0.011		
	C	0.25	0.61	0.06	0.033	0.014	0.02	0.01	<0.01	<0.01	<0.01	0.011		
	D	0.23	0.56	0.05	0.030	0.013	0.02	0.01	<0.01	<0.01	<0.01	0.011		
	AVE	0.25	0.57	0.06	0.032	0.014	0.02	0.01	<0.01	<0.01	<0.01	0.011		
											(a)			
BG2	A	0.20	0.59	0.21	0.029	0.022	0.13	0.07	0.02	0.01	0.03	0.006		
(Fully-Killed)	B1	0.20	0.57	0.21	0.029	0.017	0.14	0.08	0.04	<0.01	0.03	0.006		
	B2	0.17	0.54	0.20	0.022	0.014	0.12	0.07	0.05	0.01	0.03	0.006		
	C	0.22	0.62	0.21	0.029	0.018	0.14	0.07	0.05	0.01	0.03	0.006		
	D	0.21	0.59	0.21	0.028	0.017	0.12	0.08	0.08	0.01	0.03	0.006		
	AVE	0.20	0.58	0.21	0.027	0.018	0.13	0.07	0.05	0.01	0.03	0.006		
								(a)	(a)	(a)	(a)		(a)	(c)
BG3	A	0.22	0.55	0.20	0.028	0.013	0.12	0.07	0.04	0.02	0.04	0.006	<0.005	
	B1	0.21	0.56	0.23	0.029	0.013	0.12	0.07	0.04	0.02	0.04	0.006	<0.005	
	B2	0.20	0.55	0.20	0.026	0.013	0.12	0.07	0.04	0.02	0.04	0.006	<0.005	
	D	0.19	0.55	0.21	0.026	0.013	0.14	0.06	0.04	0.01	0.03	0.006	<0.005	
	AVE	0.20	0.55	0.21	0.027	0.013	0.12	0.07	0.04	0.02	0.04	0.006	<0.005	
BG4	A	0.23	0.55	0.22	0.034	0.017	0.12	0.10	0.07	0.01	0.02	0.006	(a) <0.01	(a) <0.01
											(c)	(c)		
BG2 C (FINE) (b)		0.23	0.56	0.21	0.031	0.023	0.13	0.07	0.05	0.02	0.031	0.007	0.015	0.014
C (COARSE)		0.24	0.55	0.21	0.033	0.022	0.13	0.07	0.05	0.02	0.030	0.007	0.016	0.012
	AVE	0.24	0.56	0.21	0.032	0.022	0.13	0.07	0.05	0.02	0.030	0.007	0.016	0.013
BG2-4	RANGE	0.17-0.24	0.55-0.62	0.20-0.23	0.022-0.034	0.013-0.023	0.12-0.14	0.07-0.10	0.02-0.08	<0.01-0.02	0.02-0.031	0.006-0.007	0.015-0.016	0.012-0.014
	AVE(d)	0.21	0.57	0.21	0.029	0.017	0.13	0.08	0.05	0.01	0.027	0.006	0.016	0.013

FOOTNOTES:

(a) Spectrographic
(b) Fine and coarse grain material taken from opposite ends of the 2½" Plate of BG2.
(c) Acid-Soluble amounts
(d) Not including spectrographic
(e) Not analyzed for

AVERAGE RESULTS OF 6 TENSILE TESTS AT 80°F. (Ref. 1)

Yield Strength (0.002-offset)	30.2 Kips/in.²
Tensile Strength	65.6 Kips/in.²
Elongation in 4 inches	37.4 per cent
Reduction of Area	53.2 per cent

Fig. 2 Fracture and defect in BG2 tested at −40° F.

The radiographs disclosed heavy porosity and slag inclusions in the through welds, scattered slag inclusions in the diaphragm fillet welds of Girder 1, and semicontinuous slag inclusions in both the through welds and the fillet welds of Girder 2. In some cases the continuity of the defects indicated a lack of fusion. Thus, the radiographs disclosed notches in the form of weld defects which may be deemed to be present to a greater or lesser extent in any large welded structure. An example of slag entrapment exposed by the fracture in BG2 is shown in Fig. 2.

A series of four, 9-ton welded structural steel girders 2½ ft. in over-all width and 2 ft. 1½ in. in over-all depth were tested, supported on cylindrical bearings with a span of 22 ft. and loaded at two points 5 ft. apart to provide a uniform bending moment over the central section. Each girder was loaded to a predetermined level, unloaded, and reloaded to higher level in successive steps to failure. Loading was continued until failure occurred or the supports prevented further bending (see Fig. 3).

A pilot specimen BG1 constructed of semikilled steel, tested at room temperature (approximately 80° F.), failed with a cleavage-type fracture after deflecting 8 in. at mid-span. Subsequent beams BG2, 3, 4 and 5 were constructed from one heat of a fully killed steel and were tested at temperatures of −40°, 0°, 40° and 80° F. The beams tested at −40° and 0° F. failed with cleavage fracture; the beams tested at 40° and 80° F. deflected until the supports prevented further bending. The preceding Table 1 summarizes the flexural test results.[2]

It may be deduced, therefore, that a small-scale test must indicate a transition temperature in the range of 0° to 40° F. in order to provide a quantitative prediction of the 9-ton box-girder performance.

TESTS AND RESULTS

Material and Location of Test Plates

Excess material remaining from the large plates used in the fabrication of the girders was forwarded to this laboratory. Identification and location of these plates with respect to their position in the fabricated girder are indicated in Fig. 4. These plates provided the initial material for the small-scale laboratory test specimens. When this material was exhausted, the broken halves of BG2 (tested at −40° F.) were used to provide additional material. This girder had deformed only slightly; hence the material several feet removed from mid-span was considered to be free of any appreciable strain effects. Tests to be described later support this supposition.

Chemical Compositions of the Steels

Chemical composition of the steel used in constructing the girders was determined for each of the several plates comprising Girders 1 and 2. Check analyses were also made on plates of Girders 3 and 4. The results are presented in Table 2. No significant variations in chemical composition were noted between plates of a single girder or between plates from different girders made of the fully killed steel.

Microstructure

Metallographic examination conducted initially by single spot checks revealed that the various 1½-in. plates of BG2 had a coarse-grained ferrite-pearlite structure with relatively little banding and with no marked nonuniformity between plates, and that the 2½-in. material from BG2 had a similar but fine-grained structure. However, an apparently anomalous variation in notch sensitivity from one location to another in the 2½-in. plate of BG2 (described later in the report)

Table 3—Standard Nick-Bend Test

Steel	*Plate thickness, in.*	*Plate grain size*	*Orientation of notch*	*Transition temperature* *Unwelded, ° F.*	*Bead welded, ° F.*
BG1	1½	Coarse	Parallel to the surface	45	135
Semikilled	1½	Coarse	Parallel	−20	40
BG2	1½	Coarse	Perpendicular	+20	40
Fully killed	2½	Fine	Parallel	< −50	30

prompted a more extensive examination of the microstructure from end to end in the $2^1/_2$-in. plate of BG2 together with a more complete sampling of the $1^1/_2$-in. material of BG2 and of the $2^1/_2$-in. plate of the other girders. The following conditions were found (see Fig. 5 for representative structures): (1) There was little difference between surface and center (with respect to thickness) of any one plate in either $1^1/_2$- or $2^1/_2$-in. material other than a tendency toward more pronounced banding at the center in banded regions; (2) one end of the $2^1/_2$-in. plate of Girder 2 was fine grained and relatively uniform; the other end was coarse and banded; (3) the fine-grain structure of the $2^1/_2$-in. plate became strongly banded approaching mid-span (the fracture was contained in this fine grain, banded region); (4) about 1 ft. beyond the fracture, in the direction of the coarse grain end, the structure was mixed with one surface of the plate coarse and the other fine and (5) the coarse structure of the $2^1/_2$-in. plate of Girder 2 was similar to that of the $1^1/_2$-in. plate material of BG2 and the $2^1/_2$-in. material of BG4 and 5. The $1^1/_2$-in. plate from Girders 4 and 5 was not investigated and the material from BG3 was not available. Thus, marked variation in microstructure occurred in the 22-ft. span of the $2^1/_2$-in. plate of BG2, which (as will be shown and discussed in a subsequent section of this report) produced appreciable differences in notch sensitivity—the fine-grain structure being superior to the coarse grain.

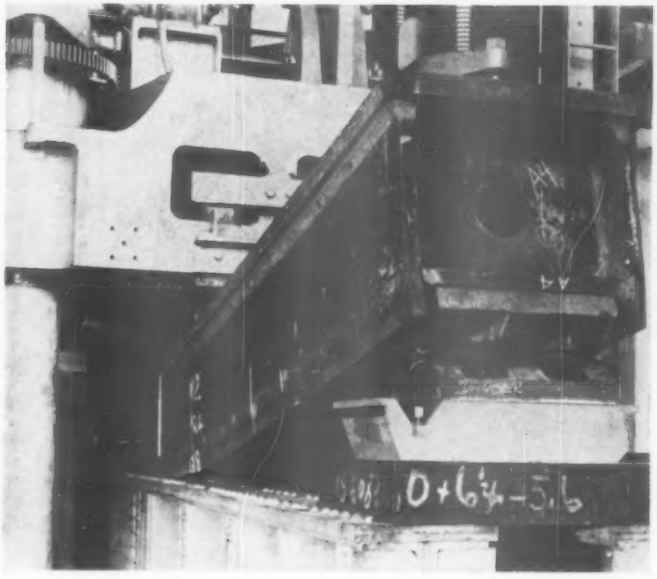
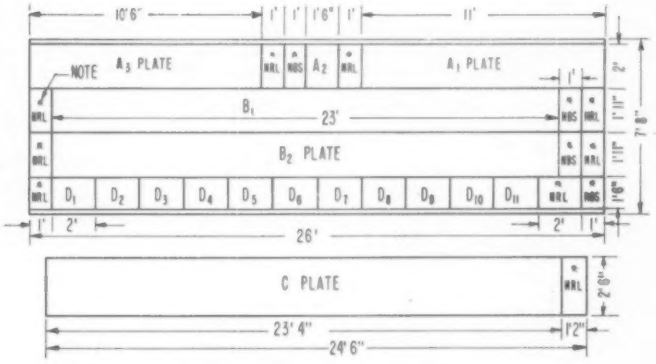


Fig. 3 Deformation in BG3 tested at 80° F.



NOTE: EXCESS PLATE MATERIAL FOR SUPPLEMENTARY SMALL SCALE TESTS
NRL - NAVAL RESEARCH LABORATORY
NBS - NATIONAL BUREAU OF STANDARDS

Fig. 4 Plate layout for box girders

Table 4—High-Constraint Nick-Bend

Steel	*Plate thickness, in.*	*Plate grain size*	*Orientation of notch*	*Transition temperature Unwelded, ° F.*	*Bead welded, ° F.*
BG2	$1^1/_2$	Coarse	Parallel to the surface	75	70
Fully killed	$2^1/_2$	Fine	Parallel	60	..
	$2^1/_2$	Coarse	Parallel	..	60

Nick-Bend Tests

Both standard and high-constraint nick-bend specimens[5] were prepared as shown in Fig. 6. The standard nick-bend specimens were notched transverse to direction of rolling and both parallel and perpendicular to the rolled surface; the high-constraint nick bends were notched transverse to rolling and parallel to the surface. A set of standards was employed for the purpose of qualitatively classifying the amount of energy expended in fracturing the specimens. The standards consisted of three major categories of load-deflection diagrams (Fig. 6). The "A-type" signified progressive failure until the load had dropped over 50%, "B-type" signified progressive failure followed by an appreciable drop in load after maximum with negligible increase in angle of bend, and "C-type" signified failure occurring at maximum load (brittle failure). Transition temperature was defined as that temperature at which there occurred a change from B- to C-type failure. The following table summarizes transition temperatures obtained with the standard nick-bend specimen.

The high constraint nick-bend specimen,[4] with its greater width and side notch, limits deformation to a smaller volume and provides a more severe state of stress than the standard nick-bend test, and, consequently, produces cleavage fractures with relatively little general plastic deformation as compared to the standard nick-bend specimen. Figure 6 presents diagrams showing typical differences between load-deflection diagrams of the standard and the high constraint nick-bend specimens in both welded and unwelded material. Table 4, at the left, summarizes the transition temperatures according to the high-constraint nick-bend specimen:

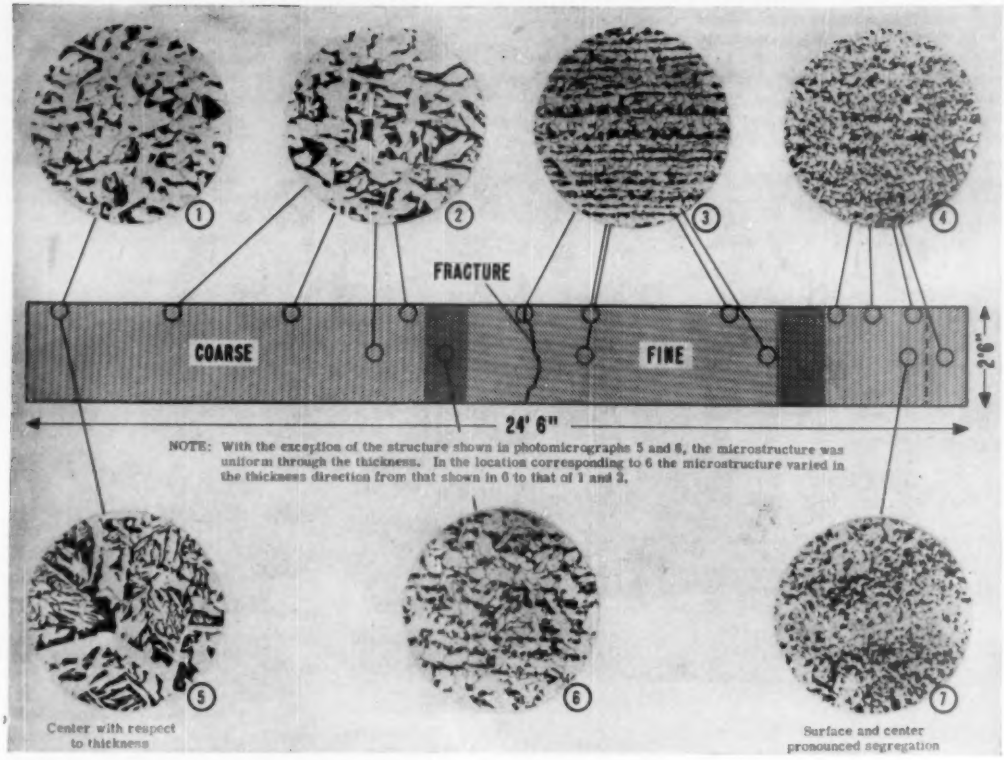

1
2
3
4
FRACTURE
COARSE
FINE
2'6"
24' 6"
NOTE: With the exception of the structure shown in photomicrographs 5 and 6, the microstructure was uniform through the thickness. In the location corresponding to 6 the microstructure varied in the thickness direction from that shown in 6 to that of 1 and 2.
5
6
7
Center with respect to thickness
Surface and center pronounced segregation

Fig. 5 Nonuniformity of microstructure in the 2½-in. plate, BC2. Mag. 100 ×, reduced 25% in reproduction

Table 5—Summary of V-Notch Charpy Transition Temperatures

MATERIAL		SPECIMEN		MICRO-STRUCTURE	UNWELDED				WELD-ANNEALED			
THICKNESS	PLATE	ORIENTATION OF NOTCH	LOCATION WITH RESPECT TO THICKNESS		ONSET OF BRITTLENESS	50% MAX. ENERGY	50% SHEAR FRACTURE	15 Ft.-lb. ENERGY LEVEL	ONSET OF BRITTLENESS	50% MAX. ENERGY	50% SHEAR FRACTURE	15 Ft.-lb. ENERGY LEVEL
BOX GIRDERS 1 (SEMI-KILLED)	A,B,C,D	PARALLEL AND PERPENDICULAR	SURFACE AND CENTER	COARSE	(Note 1) 190^{+11}	130	-	70	250^{+6} (These data apply only to notch parallel at surface)	100	150	30
BOX GIRDERS 2,3,4 and 5 (Fully-killed) 1-1/2 IN.	A	PERPENDICULAR	SURFACE	COARSE	170^{+2}	INSUFFICIENT				INSUFFICIENT		
	A	PARALLEL	SURFACE	COARSE	115^{+3}	DATA			170^{+3}	DATA		
	B1	PARALLEL	SURFACE	COARSE	(Note 2) 170^{+2} 120^{+14}	FOR SINGLE			150^{+1}	FOR SINGLE		
	B2	PARALLEL	SURFACE	COARSE	130^{+8}	PLATE POSITIONS			150^{+3}	PLATE POSITIONS		
	B2	PERPENDICULAR	SURFACE	COARSE	145^{+3}				200^{+1}			
	D	PERPENDICULAR	SURFACE	COARSE	165^{+3}							
COMPOSITE DATA OF 1-1/2" PLATES	A,B,D	PARALLEL	SURFACE	COARSE	115-170	70	65	20	150-170	50	70	-30
		PERPENDICULAR	SURFACE	COARSE	145-170	60		10	200	50	70	-10
2-1/2 IN.	C	PARALLEL	SURFACE	COARSE	200^{+3}	70	90	20	180^{+3}	50	90	-10
		PARALLEL	SURFACE	FINE	80^{+8}	20		-20	190^{+5}	50	50	-20
		PARALLEL	CENTER	FINE	80^{+5}	10		-20				
		PERPENDICULAR	SURFACE	FINE	70^{+7}	30	40	-10	160^{+6}	40	50	-30
		PERPENDICULAR	CENTER	FINE	120^{+3}	60	70	0	160^{+3}	90	60	20

Note 1:
THE PLUS FIGURE DENOTES THE NUMBER OF COMPLETELY DUCTILE FRACTURES THAT OCCURRED AT OR ABOVE THE TEMPERATURES REPORTED.

Note 2. DATA FROM TWO SEPARATE LOCATIONS IN THE SAME PLATE SUBSECTION

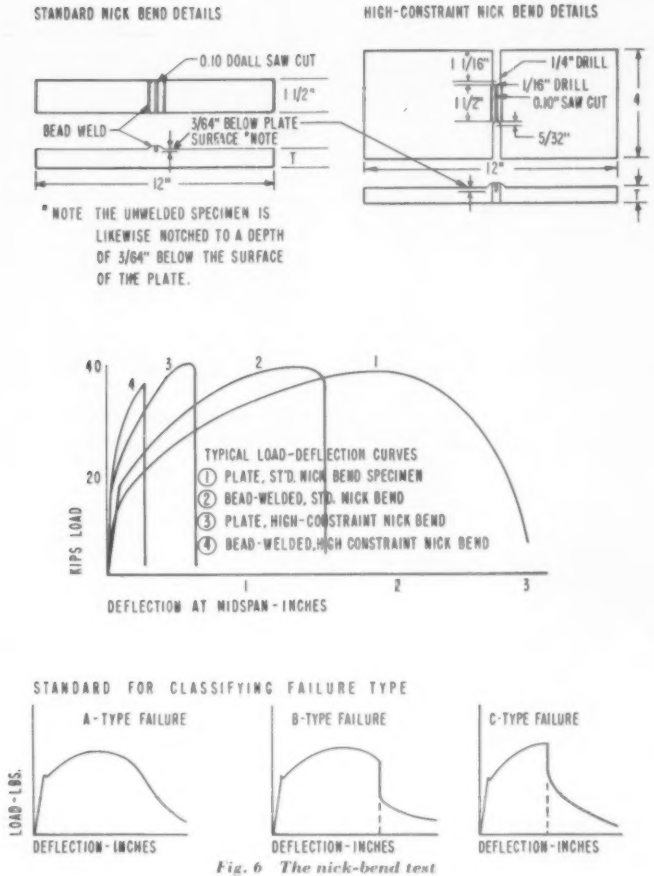


Fig. 6 The nick-bend test

V-notch Charpy Tests

The standard Charpy test was used to study variation in scatter, slope and relative position of the transition ranges of the box-girder material. In addition to the standard methods of evaluating the performance of Charpy specimens by energy absorption and fracture appearance, it was considered that useful information might be obtained by measurement of lateral contraction in the Charpy bar. It was found, however, that a strictly linear relationship existed between lateral contraction and energy absorbed in V-notch Charpy impact (see Appendix); hence this procedure was discontinued as duplicating effort.

Of the four possible major orientations of notch, two orientations were studied: (1) notched parallel to the rolled surface and (2) notched perpendicular to the rolled surface. In each case the notch was transverse to the direction of rolling. The effect of weld heat-affected zone was determined by depositing a bead transverse to the direction of rolling on 6- x 6- x $1^1/_2$-in. plates using $^5/_{16}$-in. diameter E6010 electrodes at 175 amp., 26 v. and 6 in./min. travel speed. The notch was cut in the weld metal with the apex of the notch touching the fusion line. For purposes of notching the Charpy bar perpendicular to the rolled surface, a weld bead was deposited across the edges of plates clamped together to form a composite plate 6 x 6 x $1^1/_2$ in. Specimens were taken at the surface of the $1^1/_2$-in. plate and from surface and center of the $2^1/_2$-in. plate.

Individual transition ranges were determined for each of the plates, notch orientations and surface vs. center locations. Comparisons between energy vs. temperature curves (Figs. 7–11) disclosed the following:

(1) Semikilled Steel (BG1)

The data of BG1 showed no marked difference between plates, the two orientations of notch, and center vs. surface locations (see Fig. 7 (*a*)).

(2) Fully Killed Steel (BG2)

The transition range of the fully killed steel occurred at an appreciably lower temperature level and had a higher maximum energy level than the semikilled steel (Compare Figs. 7 (*a*) and 8 (*c*)). Furthermore, there were marked differences in some cases between the two orientations of notch and the two thicknesses of plate which were not noted in the case of the semikilled steel.

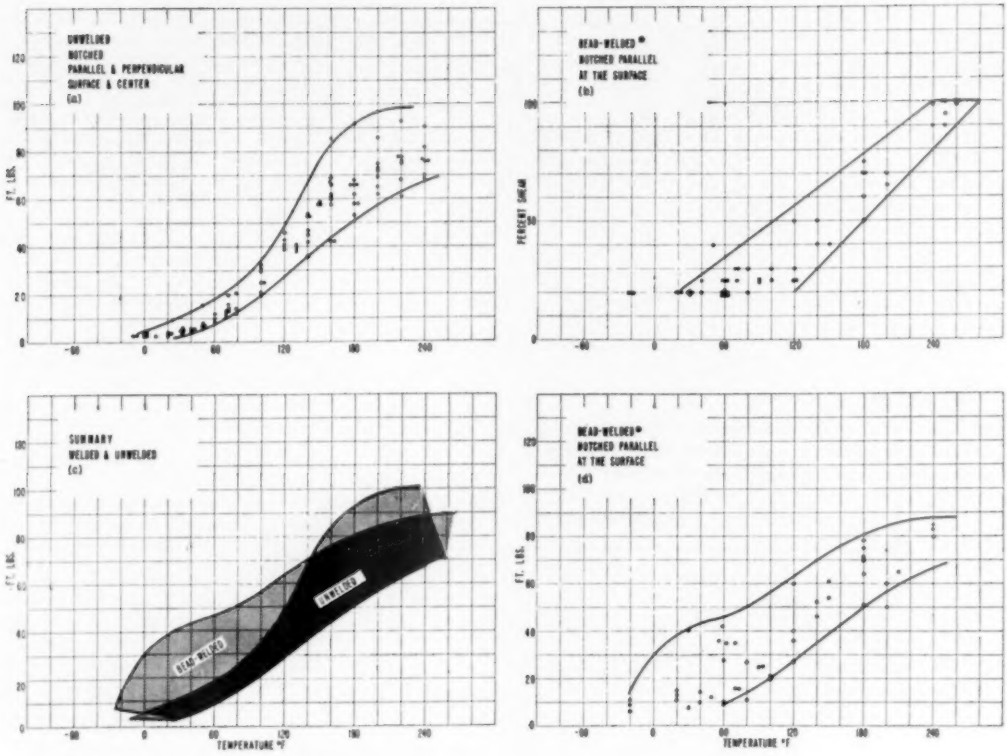
(*a*) *$1^1/_2$-in. Plate.* The transition curves of the $1^1/_2$-in. material of BG2 showed no marked difference between plates (Fig. 8 (*c*)). There was a difference, however, in scatter and in maximum energy level between notching parallel and notching perpendicular to the surface. Parallel notching produced an exceptionally wide scatter band and a higher maximum energy level.

(*b*) *$2^1/_2$-in. Plate.* The fine-grain portions of the $2^1/_2$-in. material of BG2 were decidedly superior to the coarse-grain ($1^1/_2$ in.) material of the same girder. (Compare Figs. 8 (*c*) and 10). In the case of the coarse-grain material from the $2^1/_2$-in. plate, the upper limit of the transition curve was approximately the same as that of the $1^1/_2$-in. material which was also coarse grained. The width of the scatter zone, however, was less. An appreciable difference between surface and center of the fine-grain portion of the $2^1/_2$-in. plate was indicated when notched perpendicular to the rolled surface. The transition range for the center material occurred at a higher temperature and the maximum energy level was lower than that of material taken from the surface of the plate. (Compare Figs. 10 (*c*) and 11 (*c*)). These differences were shown with perpendicular notching but not with parallel notching.

(3) Effect of Welding

(*a*) *Semikilled Material.* The notch sensitivity of the bead-weld heat-affected zone in the semikilled steel of BG1 was essentially the same as that of unwelded plate except at low-energy levels where the energy was erratic with many values as much as twice the average energy of unwelded plate at the same temperature (Fig. 7 (*c*)).

(*b*) *Fully Killed Material.* In the $1^1/_2$-in. material of BG2, welding lowered the maximum energy level and lessened the notch sensitivity in the range from 0 to −60° F. for conditions of notch parallel to the surface; little or no effect was indicated when the notch was perpendicular to the rolled surface. (Compare Figs. 8 (*e*) and 8 (*f*)). In the fine-grain $2^1/_2$-in. plate, the effect of welding varied from surface to center and with orientation of notch. In the case of perpendicular notches at the surface, welding widened the scatter zone and lowered the maximum energy value (Fig. 10 (*f*)); at the center, welding had no



*BEAD-WELDED DATA ARE NOT AVAILABLE FOR OTHER COMBINATIONS OF ORIENTATION AND POSITION.

Fig. 7 V-notch Charpy transition curves box girder 1 (semikilled). Includes 1 1/2- and 2 1/2-in. plate

effect on scatter but again lowered the maximum energy level (Fig. 11 (*e*)). With the specimens notched parallel at the surface, there was a marked widening of the scatter zone toward higher temperature (Fig. 10 (*e*)).

CORRELATION ON A BASIS OF ARBITRARILY CHOSEN CRITERIA

The primary objective of this investigation as established originally was to attempt a prediction of the performance of a 9-ton welded box girder from small-scale laboratory test specimens prepared from the girder material. It should be recalled that the girders tested at 40 and 80° F. deflected without fracturing until the supports prevented further bending, and that the girders tested at −40 and 0° F. failed with cleavage-type fracture after a small deflection. Thus, the only possible performance criterion for the box girder was the temperature at which the structure resisted fracture in bending to the limit of deflection. The various criteria and definitions of transition temperature used in the investigation of laboratory specimens are examined in the following sections for a quantitative prediction of box-girder performance. Only the 1 1/2-in. material of the fully killed steel is examined in this connection because failure in the girders initiated in the 1 1/2-in. bottom or side plates. Inasmuch as only one pilot box girder test was made with the semikilled steel, no observations relative to the transition performance of this steel are possible. Furthermore, bead-weld data are not examined for a quantitative prediction of girder performance because bead welds do not adequately simulate the complex heat effect of multipass welds as contained in the box girder. Arbitrary bead-on-plate tests were useful, however, in demonstrating the effect of welding on transition temperature in the box-girder steels. These data are discussed in a later section as complicating factors affecting correlation.

A summary of transition temperature data on 1 1/2-in. plate material from the semikilled and fully killed steels (shown in Table 5 and presented graphically in Fig. 12) demonstrates the interplay of mechanical and material variables which complicate the correlation problem. From these summaries, the following observations may be made relative to the possibilities of quantitative correlation between the performance of laboratory test specimens and the 9-ton box girders constructed from the fully killed steel:

Energy after Maximum Load Criterion (Nick-Bend)

The standard nick-bend specimen indicated a quantitative correlation with box-girder performance for the case of notch perpendicular to the plate surface (20° F.). However, with notch parallel to the surface, this correlation no longer held since the transition temperature was shifted to −20° F. Limitations due to specimen design did not permit testing of the high-constraint nick-bend specimen with notch perpendicular to the surface. Notched parallel to the surface, this specimen indicated a transition of +70° F.

Onset of Brittleness Criterion (Charpy)

Roop has recommended the use of a statistical analysis of the scatter zone and suggested that the upper limit of the zone indicates the significant transition temperature.[7] The first visual evidence of brittleness in the fracture of the V-notch

Charpy has been found to be a useful method of establishing the upper limit of the scatter zone. Although neither the plots of energy vs. temperature nor microexamination revealed significant variations from one location to another in the $1^1/_2$-in. plate material of BG2, the onset of brittleness disclosed appreciable variations. Temperatures corresponding to the onset of brittleness were found to vary from 120 to 170° F. Thus, the onset of brittleness in the V-notch Charpy specimen occurred at a temperature approximately 100° F. higher than the lowest temperature at which the girder resisted fracture.

50% Maximum Energy Criterion (Charpy)

Investigators have in some cases defined transition temperature as that corresponding to 50% of the maximum energy absorbed in fracturing the specimen. The transition temperature according to this criterion was indicated to be approximately 70° F. or about 30° higher than the lowest temperature at which the box girder resisted fracture.

50% Shear Fracture Criterion (Charpy)

When these studies were begun, there was no attempt to estimate per cent brittleness in the fracture. These data were obtained later in the investigation, however, and indicated that the transition temperatures according to this criterion correspond roughly to those indicated by the 50% maximum energy criterion. Hence, the same conclusions are applicable.

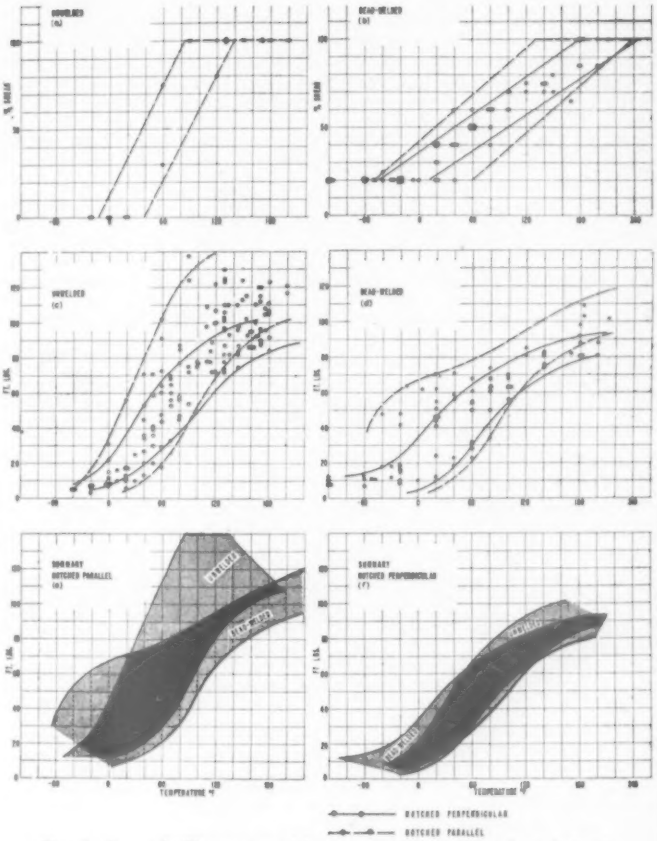


Fig. 8 V-notch Charpy transition curves $1^1/_2$-in plate, box girder 2. (Surface of the plate)

15 ft.-lb. Criterion (Charpy)

A number of investigators have advocated the use of the 15 ft.-lb. level at a specified minimum service temperature as an acceptance test for steel. An investigation at the National Bureau of Standards[8] of fractured steel plates removed from welded ships revealed a relationship between crack propagation in service and transition temperature determined by the V-notch Charpy impact test using 15 ft.-lb. as the criterion of performance. Without exception, the energy absorbed in V-notch Charpy impact was less than 15 ft.-lb. at the temperature of ship failure for plates in which fractures originated. Thus, by using this definition of transition temperature, a definite relation was found between the behavior of ship plates in service and the notch toughness of the same plates as measured by the test. It was observed that plates in which fractures ended had significantly higher energy absorption than those in which fractures started or through which cracks propagated. The temperature corresponding to 15 ft.-lb. on the faired Charpy curves for the box-girder material indicated a transition temperature of approximately 20° F. for unwelded $1^1/_2$-in. plate. Thus, by this definition the transition temperature approximated the lowest temperature at which the box girder resisted fracture.

DISCUSSION OF THE COMPLICATIONS AFFECTING CORRELATION

In the previous section it was shown that quantitative correlation was obtained in isolated cases by arbitrary selection of specimen, test procedure and performance criterion. It is improbable that such simple correlations could be repeated with other heats of even the same steel analysis. This belief is based on considerations relative to the permutation of complicating factors which determine performance. The following sections provide a critical analysis of but few of many such factors which should be considered. The examples, however, cover a sufficiently broad range to establish the principle that a rigorous, nonstatistical correlation between simple laboratory specimens and structures cannot be achieved.

Complications Involving Choice of Criterion and Definition of Transition Temperature

The selection of a transition temperature is complicated by the fact that frequently there is no well-defined critical temperature of brittleness. Instead there is a zone of scatter in which at a given temperature there is both ductile and brittle performance. The many laboratory tests in current use all exhibit this scatter to varying degrees. Moreover, there is some disagreement as to the physical significance of the scatter. This investigation has shown that the slope and width of the scatter band are dependent, at least in part, upon microstructure and orientation of notch (Figs. 8 (*c*); also Fig. 9 (*c*) vs. 10 (*c*)).

Prior to the problem of selecting some one temperature to represent the transition zone, consideration must be given the

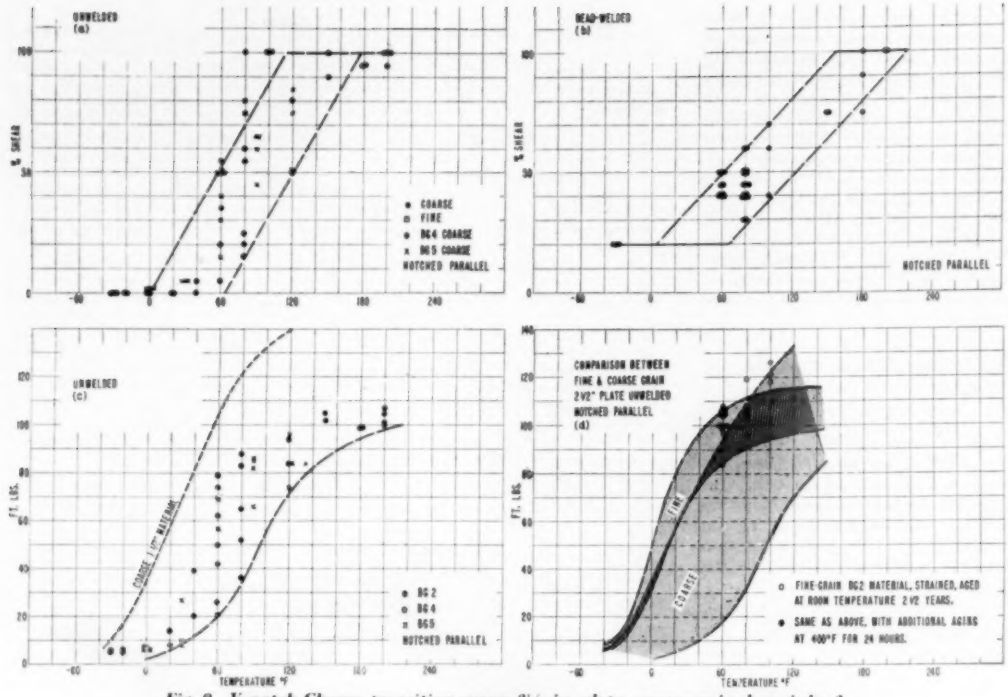


Fig. 9 V-notch Charpy transition curves 2¹/₂-in. plate, coarse grain, box girder 2. (Surface of the plate)

selection of a suitable criterion of performance. Investigations at Lehigh University[9,10] have pointed out the shortcomings of any attempt to establish correlation between tests where the transition temperatures are not all based on the same criterion. Unfortunately, it is not well understood just what properties are measured by the various criteria. The arbitrarily selected criteria and attending definitions of transition temperature examined in this investigation indicated a wide range of transition temperatures; for example, in the case of V-notch Charpy (impact) the transition temperatures ranged from approximately 0 to 200° F.

Anisotropy and Heterogeneity

The results of a study of the 2¹/₂-in. plate of BG2 provided an opportunity for an investigation of the complications due to anisotropy and heterogeneity within a single commercial plate (Fig. 13). The comprehensive Charpy survey showed a wide variation in performance (Table 6) due to changes in position of the specimen and the orientation of notch.

The wide range of transition temperatures obtained in this single plate using only one type of test specimen highlights the sampling problem which is one of the major difficulties to the general problem of correlation; i.e., the difficulty of representing laboratory performance with a simple numerical value.

Table 6—Observed Variations in Transition Temperature Resulting from Anisotropy and Heterogeneity Effects in a Single As-Rolled Plate

(C-Plate BG2 2¹/₂ in. x 2 ft. 6 in. x 24 ft. 6 in.)

	Same position in plate different notch orientations (anisotropy effect), ° F.	*Same orientation of notch different plate positions, (heterogeneity effect)*, ° F.
Onset of brittleness	80/120	80/200
50% Max. energy	10/60	20/70
15 Ft.-lb. energy	−20/0	−20/20

Rate of Loading

Klier, Wagner and Gensamer[11] found a marked lowering of the transition temperature (according to the 50% maximum-energy criterion) as the result of changing from conventional V-notch Charpy "impact" to slow bend. A similar study was made using fine- and coarse-grain material from the 2¹/₂-in. fully killed box-girder material. The bend jig incorporated the same dimensions of distance between supports, radii at the supports and radius of plunger as in standard Charpy impact test equipment. The temperature corresponding to 50% maximum energy was noted from the energy-temperature curves, and the temperatures corresponding to the onset of brittleness and 50% shear were noted from the fractures (Fig. 14).

From Table 7, it may be seen that all criteria indicated speed effects but to varying degrees. The criteria based on fracture appearance indicated a marked lowering of the transition temperature as the result of the decrease in speed from impact to slow bend.

Table 7—Summary of Transition Temperature (° F.) V-Notch Charpy

Slow Bend (14 x 10⁻⁶ ft./sec.) Versus Impact (17 ft./sec.)

	Fine-grain material		*Coarse-grain material*	
Criteria	*Impact*	*Slow bend*	*Impact*	*Slow bend*
50% Max. energy	20	−10	70	50
50% Shear fracture	..	−10	90	30
Onset of brittleness	80	20	200	90

Strain Aging

The question of strain aging has risen repeatedly throughout this investigation in connection with the lapse of time between the straining of the girders and the testing of the small-scale laboratory specimens. Moreover, due to the intermittent loading of the girders, strain aging during the process of testing was possible — at least in the girder tested at 80° F.

Two simple methods were employed to determine the strain-aging propensities of the box-girder steel. The first method consisted of an evaluation of the aging effect which occurred near the fracture of BG2 (tested at −40° F.) after a lapse of about $2^1/_2$ years. Twelve Charpy specimens were taken from material within 1 ft. of the fracture. Six of the specimens were tested in the room-temperature aged condition, and six were given an additional artificial aging treatment of 400° F. for 24 hr. The specimens were tested at a temperature just above the transition range (60° F.) and the data were compared with those from unstrained plate material which *had* been set aside in the fabrication of the girder. No differences in notch sensitivity were observed; i.e., energy values from the strained material which had aged at room temperature since the testing of the girder (approximately $2^1/_2$ yr.) and from the same material given an artificial aging treatment were essentially the same as the values from unstrained plate material (see Figure 9 (*d*)). The second method consisted of prestraining a series of high-constraint nick-bend specimens to a certain per cent of the mid-span deflection reached at maximum load, and then artificially aging. The bending before and after aging was done at a single temperature just above the transition range, viz., 80° F. At this temperature the average deflection at maximum load was 0.78 in. for the coarse-grain material of BG2. Four specimens were prestrained by bending to approximately 25% of this deflection, and four to approximately 50%. Loading was interrupted after the prestraining operation and the specimens were artificially aged at 200° F. for $7^1/_2$ hr. Testing was resumed 24 hr. after prestraining. As the result of the aging treatment, there was approximately a 1000 lb. increase (6.5% rise) in the reloading yield point for both amounts of prestraining. Had the steel not suffered aging, upon reloading the yield-point load would have corresponded to the load at which bending was interrupted. In addition to the rise in yield-point load, there was a 10% decrease in the energy after maximum load as the result of prestraining 25% and aging, and an average decrease of 75% after prestraining 50% and aging.

Welding

Welding must necessarily be considered in any attempt to establish correlation between laboratory specimens and welded structures in service. That this is an added complication is evident from the varied response of the several criteria to the introduction of a bead-weld heat-affected zone (Figs. 12 and 13). For example, onset of brittleness in V-notch Charpy impact indicated welding to have a marked effect on the semikilled steel of BG1 and also on the $2^1/_2$-in. fine-grain material of BG2 (welding raised the transition temperature 60 and 100° F., respectively); in other instances, notably the case of coarse-grain material from the $2^1/_2$-in. plate of BG2, the onset of brittleness indicated welding to have little or no effect on transition temperature. Using the 50% maximum energy criterion, welding lowered the transition temperature 20° F. in the $1^1/_2$- and $2^1/_2$-in. coarse grain material, while it raised it approximately 20° F. in the $2^1/_2$-in. fine-grain material. By the 15 ft.-lb. criterion, welding was indicated to consistently lower the transition temperature (approximately 30° F.) in the $1^1/_2$- and $2^1/_2$-in. coarse-grain material of BG2; however, in the fine-grain material of the $2^1/_2$-in. plate of BG2 this criterion indicated a varied response depending upon orientation of notch and location of specimen with respect to thickness.

In the case of the standard nick-bend specimen, welding not only appreciably raised the transition temperature but the effect completely overshadowed such variables as notch orientation and grain size to produce approximately the same transition temperature after welding for all cases (40° F.). In the case of the high-constraint nick-bend specimen the transition temperature (70° F.) was essentially unaffected by welding.

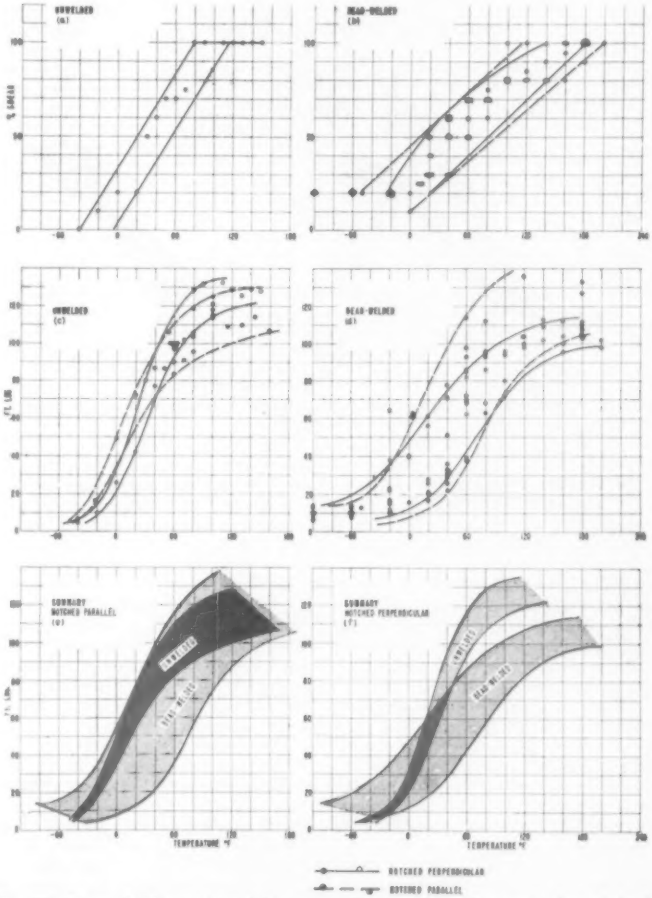


Fig. 10 V-notch Charpy transition curves $2^1/_2$-in. plate, fine grain, box girder 2. (Surface of the plate)

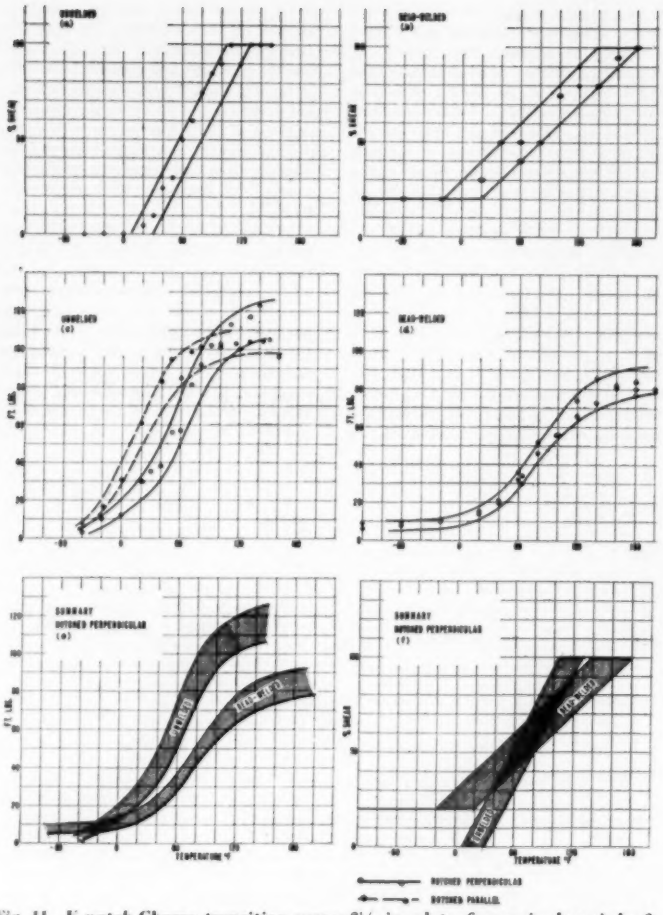


Fig. 11 V-notch Charpy transition curves 2½-in. plate, fine grain, box girder 2. (Center of thickness)

SUMMARY DISCUSSION AND CONCLUSIONS

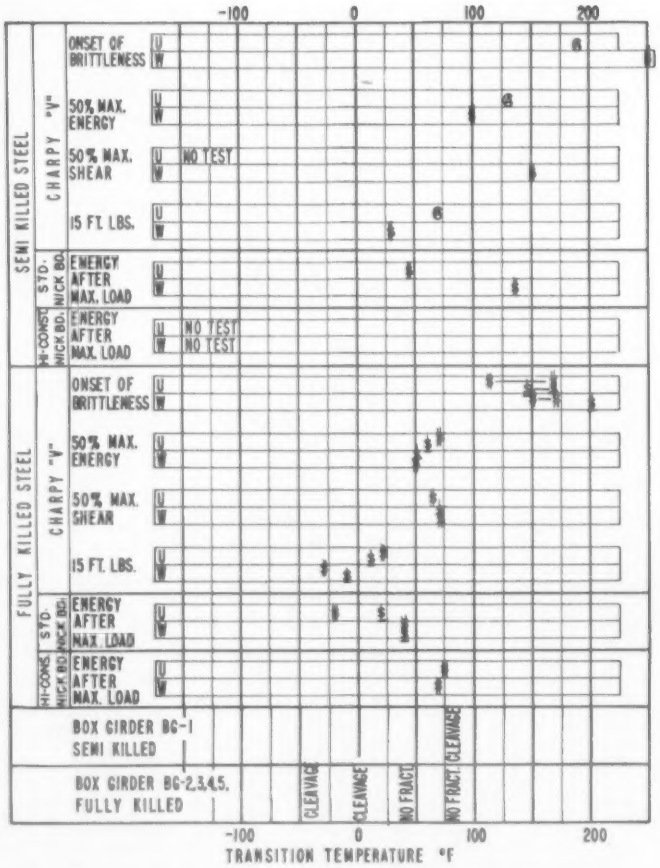
It is generally recognized that the performance of a structure or a test specimen depends on the combined effects of mechanical and material factors. Mechanical factors include the imposed conditions of service and design which determine the state of stress, rate of loading, temperature, etc. Material factors include the numerous and sometimes elusive elements which characterize steel quality, viz., notch and strain-rate sensitivity, strain-aging susceptibility, etc. The complexity of this interplay of factors is evidenced by the individuality of performance shown by the various specimens included in this investigation.

Coupled with individuality of test specimens is the added complication of choice of performance criterion. Unfortunately, it is not well understood just what properties are measured by the various criteria. It has been observed that not all criteria measure the same effects. For example the criterion of 50% maximum energy was relatively insensitive to a change from impact to slow-bend testing of V-notch Charpy bars, while onset of brittleness indicated a marked velocity effect. The transition temperature was lowered approximately 100° F. according to the onset of brittleness although it was lowered only 20° F. according to the energy criterion.

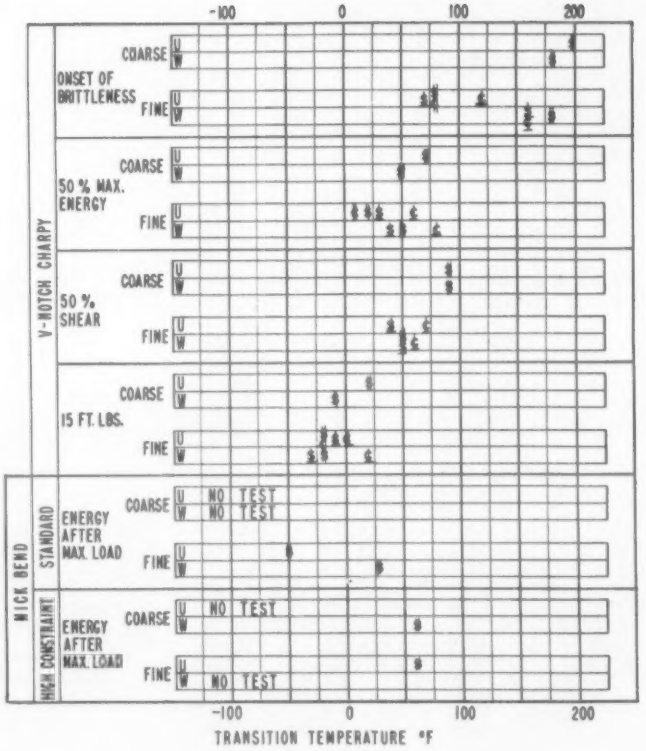
The complications arising from the interdependence of mechanical and material factors must be considered to exist even in the case of a completely homogeneous and isotropic material. Real materials, however, are neither homogeneous nor isotropic, but vary widely in their material properties from point to point. Chance fluctuations in performance should therefore be expected to a degree which is dependent on the specific sensitivity of the test specimen to the various material properties. Thus, for real materials, correlation between specimens or between specimens and structures should be deemed feasible only in a very approximate sense, and subject to wide chance fluctuations requiring statistical analysis. The problem here is one of "sampling" in a true statistical sense of the term.

The phase of this investigation which deals with effects of anisotropy and heterogeneity in a single plate provides substantive evidence for the above conclusion. V-notch Charpy data disclosed differences in transition temperature of as much as 120° F. in a single plate due to changes in sampling loci, and as much as 50° F. due to variation in orientation of notch. A change in speed of testing from Charpy impact to slow bend disclosed a marked effect that varied with grain size: slow bending lowered the transition temperature (onset of brittleness) 110° F. in coarse-grain material but only 60° F. in fine-grain material. Welding introduced an extreme condition of anisotropy and heterogeneity but its effect was found to depend on the orientation of notch, location of specimen with respect to thickness, and the grain size of the parent metal. For example, welding raised the transition temperature (onset of brittleness in V-notch Charpy impact) 100° F. in fine-grain plate material, while coarse-grain material from the same rolled plate was indicated to be relatively unaffected by welding. In the case of standard nick-bend, welding not only appreciably raised the transition temperature (based on energy after maximum load) but its effect eliminated differences resulting from such variables as notch orientation and grain size of the parent metal. In contrast, the high-constraint nick-bend specimen indicated little or no effect from welding.

Thus, the results of this investigation emphasize that the concept of a rigorous definitive correlation between the performance of a simple laboratory specimen and full-scale structure is not feasible. In the case of the box girder a fortuitous correlation might have been obtained using any of the varied specimens and test procedures investigated by a manipulation of criteria and definition of transition temperature. The various specimens, criteria and attending definitions indicated transitions ranging from approximately −80 to 200° F. for the girder material while the box girder itself resisted fracture at 40° F. Although a reasonable prediction of box-girder performance was indicated by the standard nick-bend specimen using energy after maximum load as the criterion and also by the V-notch Charpy impact specimen using 15 ft.-lb. as the criterion, there is no justification in assuming that other steels fabricated to form the same



KEY: (U) UNWELDED - PARENT PLATE (C) CENTER (WITH RESPECT TO THICKNESS)
(W) WELDED PLATE (H) NOTCH PARALLEL TO THE ROLLED SURFACE
(S) SURFACE OF THE PLATE (⊥) NOTCH PERPENDICULAR
(A) ALL CONDITIONS - NOTCH AT THE SURFACE AND CENTER, PARALLEL AND PERPENDICULAR.



KEY: (U) UNWELDED - PARENT PLATE (C) CENTER (WITH RESPECT TO THICKNESS)
(W) WELDED PLATE (H) NOTCH PARALLEL TO THE ROLLED SURFACE
(S) SURFACE OF THE PLATE (⊥) NOTCH PERPENDICULAR

Fig. 13 (above) Summary of transition temperature data on a single 2½-in. fully killed plate

Showing the effects of heterogeneity, choice of test specimen and choice of performance criterion on transition temperature

Fig. 12 (left) Summary of transition temperature data on 1½-in. plates

Showing the extent to which heterogeneity, choice of test specimen and choice of performance criterion complicate correlation

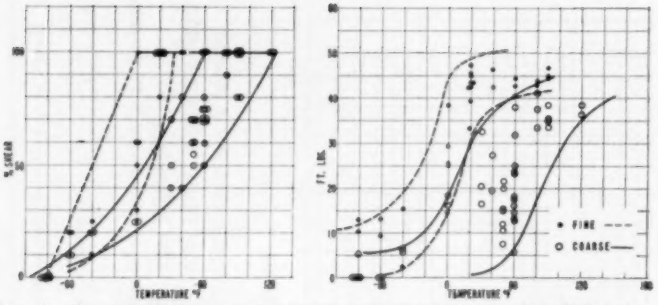
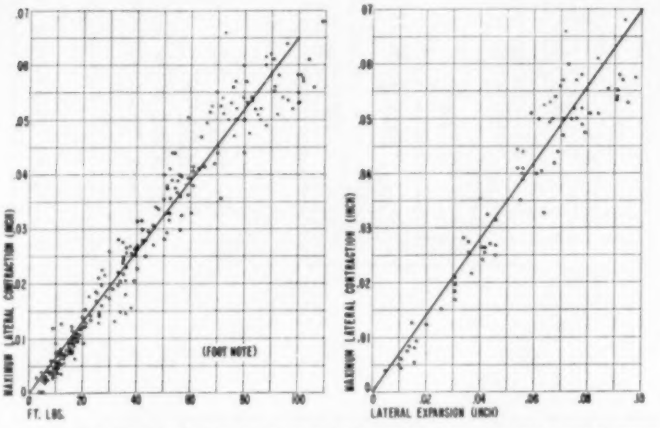


Fig. 14 V-notch Charpy transition curves—slowbend; $2^1/_2$-in. fully killed plate, unwelded, notch parallel at the surface



FOOTNOTE: THESE DATA INCLUDE SUCH CONDITIONS AS WELDED AND UNWELDED COARSE GRAIN 1-1/2 AND 2-1/2-INCH BG2 MATERIAL NOTCHED PARALLEL, UNWELDED FINE-GRAIN 2-1/2-INCH BG2 MATERIAL NOTCHED PERPENDICULAR, AND WELDED 1-1/2 AND 2-1/2-INCH BG1 MATERIAL NOTCHED PARALLEL.

Fig. 15 Lateral expansion and contraction in V-notch Charpy impact specimens

structure, or the same steel to form a similar structure, would correlate in the same empirical fashion.

The need at this time is for a basic extension of our knowledge background relative to the behavior of metals under conditions unfavorable to plastic deformation. The two factors, mechanics and material, which together determine performance are parametrically related and should be investigated in strict recognition of this fact.

ACKNOWLEDGMENTS

The assistance in the initial stages of this investigation of former members of the Metallurgy Division, viz., C. E. Jackson, F. H. Laxar, R. E. Metius, W. E. Ellis and G. G. Luther is gratefully acknowledged. The authors are indebted to Dr. O. T. Marzke, Superintendent of the Metallurgy Division for his interest and advice during the course of the investigation and for his constructive review of the final report.

APPENDIX

Lateral Contraction in Charpy Specimens

Lateral contraction has been suggested by Kinzel[12] as a measure of ductility. The specimen (3 in. wide by 8 in. long tested in slow bend) is longitudinally welded and has an Izod notch of 0.01-in. vertex radius milled transverse to the weld bead to a depth of 0.050 inch. The width of the specimen is measured with a micrometer caliper $^1/_{32}$ in. below the notch. The reduction in width is termed lateral contraction.

It was considered that useful information might be obtained in the investigation of box-girder steels by the measurement of lateral contraction in V-notch Charpy impact specimens. A jig featuring a dial gage calibrated to 0.0001 in. was constructed to measure the lateral contraction $^1/_{32}$ in. below the apex of the V-notch. At first readings were confined to the contraction on the center line directly under the notch but this was found to produce excessive scatter in the data of specimens absorbing over 50 ft.-lb. With brittle fractures, maximum lateral contraction occurred directly under the notch; but with ductile breaks the fracture path was erratic, and frequently the maximum lateral contraction was found to occur at a point considerably removed from the center line of the notch. Thus, measurements on the center line of the notch were abandoned in favor of maximum lateral contraction. Contractions were measured on both sides of each half of the broken specimen. The total contraction was taken as one-half the sum of the four readings. Where tears or jagged lips precluded four measurements, lateral contraction was taken as the sum of two readings.

The plot of maximum lateral contraction vs. energy for Charpy bars taken from $1^1/_2$-in. plate Girders 2 and 4 revealed a simple straightline relationship, independent of both testing temperature and notch orientation (Fig. 15). Furthermore the relationship held for the semi-killed steel of Girder 1 and for bead-weld heat-affected zone in both BG1 and BG2.

It was felt that the lateral expansion occurring on the compression side of the bar could be measured more easily and might provide the same simple relationship as lateral contraction. The supposition proved correct. The measurement was made in a matter of seconds with ordinary micrometer calipers and the plot of lateral expansion vs. energy absorbed was again found to indicate a simple linear relationship.

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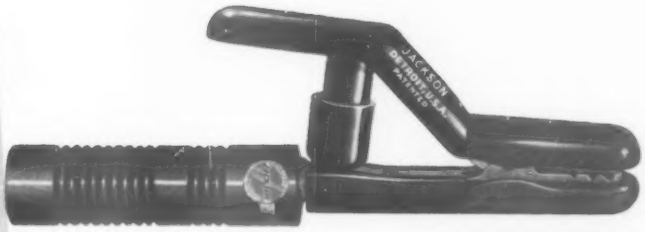

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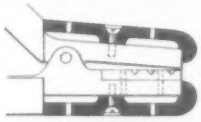

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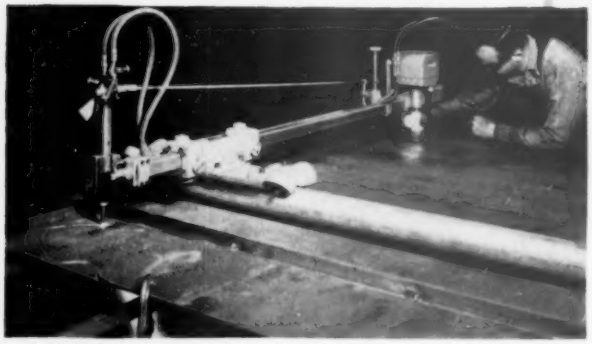

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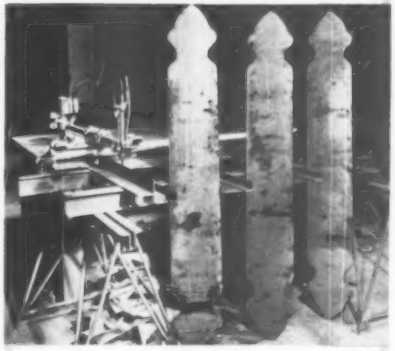

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